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SOME NOTES ON THE BEHAVIOR OF CARBON TOOL STEEL ON QUENCHING

By G. V. Luerssen

Abstract

This paper is a plea for uniform carbon tool steel. It emphasizes the necessity of uniformity through a series of illustrations showing several characteristic conditions of quenching range and hardness penetration which may exist. A convenient type of Metcalf test to determine the degree of uniformity is described. The relationship of quenching range and hardness penetration to the occurrence of soft spots, change of size in hardening and susceptibility to cracking is shown, and the influence of these characteristics upon actual tool performance is illustrated by several typical examples. The paper further emphasizes the fact that it is necessary to use a tool steel having uniform hardening characteristics if uniform behavior in quenching and in tool service is to be obtained.

CARBON tool steel, while perhaps the oldest steel known to the arts, has probably been the subject of less discussion than any other steel of like importance. While individuals handling tool steels undoubtedly have accumulated valuable information regarding its properties, behavior, etc., still the published literature contains little on the subject. This may indicate lack of sufficient understanding of the principles involved to coordinate these with the facts observed, but the writer prefers to believe that it is due to two things, first the natural feeling that the field of research in carbon steels has long been exhausted, and second the recent trend toward alloy tool steels.

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The substitution of alloys for plain carbon steels has without question solved many vexing tool problems, but there have undoubtedly been many cases in which alloys have been resorted to needlessly. Such cases are often those in which the carbon steels had produced remarkably fine tools sometimes, while at other times had fallen down badly. The problem in such cases of obtaining uniformly good tools has been attacked by going into alloys, when it might very easily have been solved by securing a uniform, high-quality carbon steel. Futhermore there are many jobs in which carbon steel has been found superior to alloy steels. These are represented notably by the class of tools which require a very hard surface to withstand wear or battering, and a tough interior to withstand splitting. Such conditions are regularly encountered in heading dies, certain types of shear blades, etc.

Aside from the fact that carbon steels are eminently adapted to certain classes of service, they are often very desirable from the point of view of the tool maker. They are more easily forged than most alloys, easier to anneal, easier to machine, and simple to harden. In many cases they are cheaper, and more readily obtainable from stock. With these many things to recommend their use, there appears to be no question that the improvement of carbon tool steels is a subject worthy of further study.

Variations in Carbon Tool Steel

The writer feels sure that every observing user of carbon tool steel has at times noticed variations in its behavior. For many years the workers of steel both in Europe and in this country have used the term "body" in describing certain characteristics of tool steel. While this term "body" has always been more or less a will-of-the-wisp, still it undoubtedly has some meaning, and probably describes some more or less involved condition. Recently Maurer and Haufe[1, 2] have studied the matter of tool steel quality, and have interpreted body as the ability of the steel to be hardened repeatedly without cracking. Egeberg[3] has reported important variations in

[1]E. Maurer and W. Haufe, "Influence on the Hardening of Tool Steels of Elements Usually Considered Detrimental." *Stahl und Eisen*, Vol. 44, 1924, p. 1720.

[2]W. Haufe, "Effect of Silicon, Nickel, Chromium and Tungsten on the Hardening Properties of Tool Steel." *Stahl und Eisen*, Vol. 47, 1927, p. 1365.

[3]Birger Egeberg, "On Electric Steel Melting," Transactions, American Society for Steel Treating, Vol. 10, September, 1926, p. 395.

the performance of carbon rock drill steels, tracing these back to certain conditions of refining in the electric furnace.

Wright[4] has studied the effect of normality and of small percentages of impurities upon the performance of carbon steel in ball heading dies, and has found an abnormal steel to show a slight advantage over the more normal samples tested.

Occasionally complaints are made that it is impossible to harden tools from certain bars of steel without soft spots. These are commonly referred to as fugitive soft spots for the reason that upon requenching they occasionally disappear, but re-occur in some other portion of the tool or part, indicating that they are not associated with any specific surface condition, but rather result from some inherent property of either the quenching bath or the steel itself. That the character of the quenching bath has a profound effect upon this condition has been shown by Epstein and Rawdon[5] who were able to produce soft spots at will on case hardened specimens by increasing the gas content of the quenching water. It is undoubtedly true however that some carbon tool steels do tend to develop soft spots in quenching under conditions which do not produce them in other steels of the same composition, and while it is true that in practically every case soft spots can be prevented by proper quenching methods, the condition mentioned above represents a variable in the steel which is not desirable.

Steels have been reported on the other hand which tend to crack in water hardening more readily than others. These often have a dry stony fracture even when quenched from as low a temperature as 1430 degrees Fahr. (775 degrees Cent.). Still other steels are reported to check in grinding more readily than others. Here again undoubtedly there are two variables to consider, viz., the character of the wheel and manner of grinding, and the properties of the steel. It is well known that any piece of hardened steel can be checked by abuse in grinding. But on the other hand it is true that some carbon steel tools check more readily than others under the same grinding conditions.

Lastly it is reported occasionally that some steels change size in hardening differently than other steels of the same composition.

[4]F. L. Wright, "High Temperature Treatment Applied to Cold Heading Ball Dies of Plain Carbon Tool Steel." TRANSACTIONS, American Society for Steel Treating, Vol. 13, February, 1928, p. 282.

[5]S. Epstein and H. S. Rawdon, "Steel for Case-Hardening—Normal and Abnormal Steel." Research Paper No. 14, Bureau of Standards, September, 1928.

While size change may often be traced back to variations in the annealed structure, it is perfectly possible to have these variations in steels annealed to the same structure, and again there appears to be some inherent property of the steel responsible for the variation.

In the course of a research covering about five years conducted for the purpose of developing methods for the manufacture of uniform carbon tool steel, studies were made of all the peculiarities mentioned above, and it was found that within certain limits these are dependent upon two properties of the steel, viz., hardness penetration and quenching range. It was found further that when these two properties were controlled within close limits, the steel gave uniform performance in service. It is not within the scope of this paper to touch upon methods of manufacture. Suffice it to say that the properties to be described depend largely upon conditions in the melting furnace, and that in order to keep these properties uniform from heat to heat, it is necessary to duplicate in exactly the same way, and time after time, a great many details of the melting operation. The paper rather will attempt to show how the peculiarities mentioned above are related to the two inherent properties, hardness penetration and quenching range, pointing out the necessity of using uniform tool steel if uniformly satisfactory results in service are to be obtained.

Standard Test for Hardening Properties

It is rather generally supposed that hardness penetration in quenched carbon tool steel is dependent entirely upon analysis, i. e., with regard to the elements usually determined in steel. This is only partly true. Chromium present as an impurity over 0.15 per cent and silicon over 0.45 per cent will of course affect penetration. Carbon content has an effect, all other conditions remaining constant, an 0.85 per cent carbon steel hardening deeper than the higher carbon steels. Manganese has an effect when present in quantities over about 0.40 per cent, and this effect seems to be magnified in the presence of higher silicon.

Taking however the limits of analysis usually set for carbon tool steel, viz., manganese (maximum) 0.40 per cent, silicon 0.10 to 0.25 per cent, chromium (maximum) 0.10 per cent, nickel (maximum) 0.15 per cent, close observation over a number of years has shown that steels of almost identical analysis may have totally different

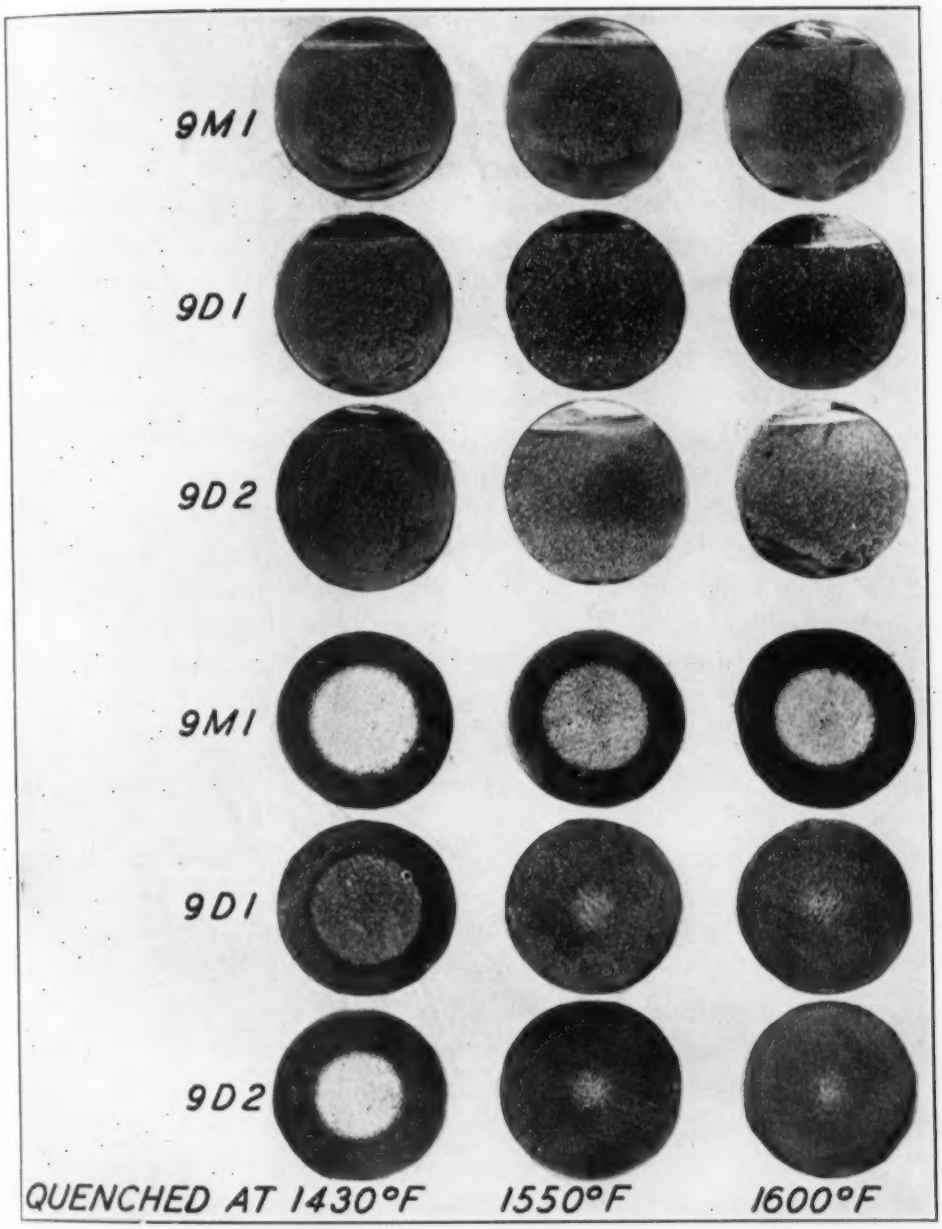

Fig. 1—Sections of Three Steels of Approximately the same Analysis Quenched from Three Different Temperatures in Brine.

The upper group shows actual fractures. The lower shows corresponding sections etched to bring out hardness penetration. Note the coarseness of fracture and deep penetration in 9D1 and 9D2 as compared with 9M1. Fracture made on ¾-inch diameter bars. Actual size.

properties with regard to hardness penetration, some hardening only to a depth of $\frac{1}{32}$ inch in a ¾-inch section, others hardening to a

depth of $\frac{3}{16}$ inch in the same section when quenched in water at 1450 degrees Fahr. (790 degrees Cent.).

The other variable, viz., quenching range is likewise not affected by analysis within the limits already prescribed. Given the same chemical analysis in two steels, one may allow of quenching in brine at temperatures up to 1650 degrees Fahr. (900 degrees Cent.) without serious coarsening of the grain, and without greatly deepening the hard shell, while the other will exhibit a coarse, dry grain when quenched at 1450 degrees Fahr. (790 degrees Cent.) and at 1550 degrees Fahr. (845 degrees Cent.) will have a very brittle, coarse fracture. It might be added that steels of the latter class are always deep hardening. Those of the former class may be either deep or shallow hardening.

The conditions described above are illustrated in Figs. 1 to 7 inclusive. Fig. 1 shows a series of sections made on ¾-inch bars from three different heats of approximately the same analysis, quenched in brine at 1430 degrees Fahr. (775 degrees Cent.), 1550 degrees Fahr. (845 degrees Cent.), and 1600 degrees Fahr. (870 degrees Cent.). These, together with all other steels shown in this paper are electric furnace steels. The sections on the upper half of the page are actual fractures. Those below are corresponding sections, etched in hot 1:1 HCl to bring out clearly their hardness penetration. It should be noted that these pieces, as well as all subsequent fractured pieces shown were quenched in a 15 per cent brine solution, the temperature of which was held between 65 and 68 degrees Fahr. (18 and 20 degrees Cent.). The pieces were agitated in quenching, but not flushed. Bars were all annealed to a spheroidized structure and were finish-machined so as to remove all possible decarburization from the bar surface and at the same time eliminate any surface variables which might affect the quench.

A marked variation in quenching characteristics among these three heats is quite apparent. Considering sample 9M1, it will be seen that when quenched at 1430 degrees Fahr. (775 degrees Cent.) this bar shows a fine silky outer shell about ⅛ inch in depth. After a 1550-degree Fahr. (845 degrees Cent.) quench, the depth of shell has increased only slightly, and the fracture has not coarsened. At 1600 degrees Fahr. (870 degrees Cent.) the penetration has increased somewhat, but the fracture has not coarsened appreciably, still remaining silky and tough after this very high temperature quench.

Table I
Analyses of Steels

Steel No.	Per Cent C	Mn	Si	P	S	Cr	Ni
9M1	0.91	0.25	0.22	0.010	0.016	0.07	0.08
9D1	0.95	0.30	0.15	0.015	0.017	0.05	0.08
9D2	0.89	0.24	0.11	0.028	0.018	nil	0.01
95M	0.95	0.33	0.16	0.015	0.020	0.06	0.11
95D	0.96	0.37	0.25	0.022	0.014	0.07	0.07
105A	1.04	0.25	0.10	0.010	0.022	0.08	0.10
105B	1.06	0.25	0.16	0.010	0.020	0.07	0.10
105C	1.00	0.38	0.20	0.022	0.012	0.08	0.11
105D	1.07	0.39	0.25	0.021	0.010	0.08	0.10
105E	1.07	0.30	0.18	0.016	0.017	0.06	0.09
9G	0.93	0.31	0.24	0.019	0.014	0.07	0.08
9P	0.90	0.26	0.18	0.018	0.019	0.07	0.08
14G	1.42	0.39	0.12	0.024	0.008	0.10	0.08
14P	1.36	0.40	0.26	0.028	0.007	0.09	0.12
10M	0.98	0.26	0.26	0.020	0.006	0.08	0.10
10D	0.94	0.27	0.11	0.023	0.005	0.10	0.12

Sample 9D1, on the other hand, shows a rather dry fracture after quenching at 1430 degrees Fahr. (775 degrees Cent.), with a penetration of about $\frac{5}{32}$ inch. Raising the quenching temperature to 1550 degrees Fahr. (845 degrees Cent.) badly overheats the steel and results in a very coarse, brittle structure and extremely deep penetration. A quench from 1600 degrees Fahr. (870 degrees Cent.) naturally has the same effect. Sample 9D2 shows an unusual penetration of hardness at all temperatures. At 1430 degrees Fahr. (775 degrees Cent.) it shows a penetration of $\frac{3}{16}$ inch with a somewhat dry, stony fracture, while at 1550 and 1600 degrees Fahr. (845 and 870 degrees Cent.) it exhibits much the same properties as 9D1; except that the fracture is not quite so coarse.

To bring out in a more striking manner the wide difference between these steels, it might be pointed out that 9M1 quenched at 1600 degrees Fahr. (870 degrees Cent.) shows a finer fracture than does 9D1 quenched at 1430 degrees Fahr (775 degrees Cent.), and a penetration when quenched at 1600 degrees Fahr. (870 degrees Cent.) shallower than that of 9D2 quenched at 1430 degrees Fahr. (775 degrees Cent.). In other words while 9M1 might be quenched at temperatures between 1550 and 1600 degrees Fahr. (845 and 870 degrees Cent.) without overheating, 9D1 and 9D2 are actually slightly overheated at a temperature of 1430 degrees Fahr. (775 degrees Cent.). Both 9D1 and 9D2 are very sensitive to overheat, possessing an extremely narrow quenching range, while 9M1 has an extremely wide quenching range and very little tendency to overheat. Analyses of the three steels are shown in Table I.

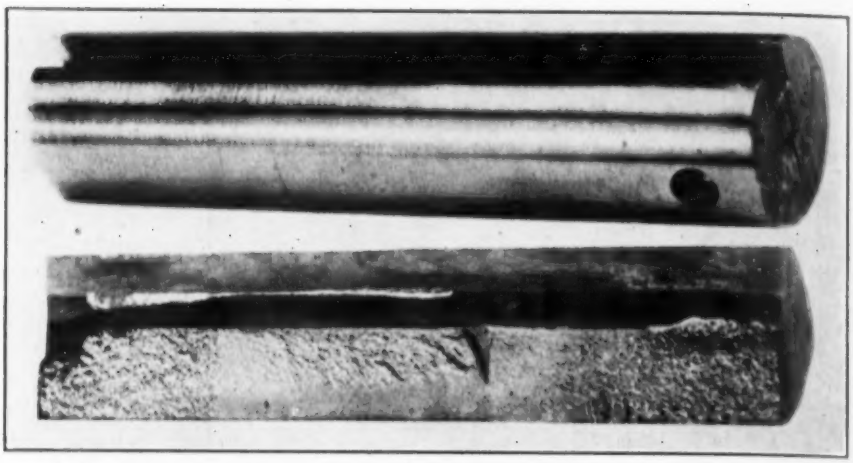

Fig. 2—Test Specimen for Showing Quenching Range and Penetration on One Fracture.
The specimen is quenched on a tapering heat and fractured through the dotted line. Actual size.

The characteristics illustrated in Fig. 1 can very conveniently be shown on one specimen by heating the bar on a tapering heat, quenching and splitting lengthwise. This will be recognized as a modification of Metcalf's test in which he used a bar containing transverse notches, fracturing into several pieces after the quench. On the type of test described here, the resulting fracture gives a complete picture of both hardness penetration and quenching range. The test as used for a number of years by the laboratory with which the writer is connected is conducted as follows:

A bar 3¼ inches long and ¾ inches in diameter is machined on the lathe, after which a longitudinal slot $\frac{3}{16}$ inches wide and $\frac{3}{16}$ inches deep is cut on one side. A $\frac{3}{16}$-inch hole $\frac{3}{16}$ inch deep is then drilled about ¼ inch from one end as shown in Fig. 2. This bar is heated in a specially constructed tube furnace in which the winding is so arranged that a gradient in temperature is obtained with the zone occupied by the bar. One thermocouple, that recording the high temperature end, is cemented into the hole in the specimen, and is introduced with it into the furnace. Two additional thermocouples equally spaced enter the furnace through the tube wall, the beads of these couples engaging the slot in the specimen. The slot is filled with alundum so as to prevent high readings due to radiation from the tube walls. The thermocouples are made of No. 16 iron and constantan wire, and are connected to a recorder.

By means of suitable rheostats the furnace is controlled so as

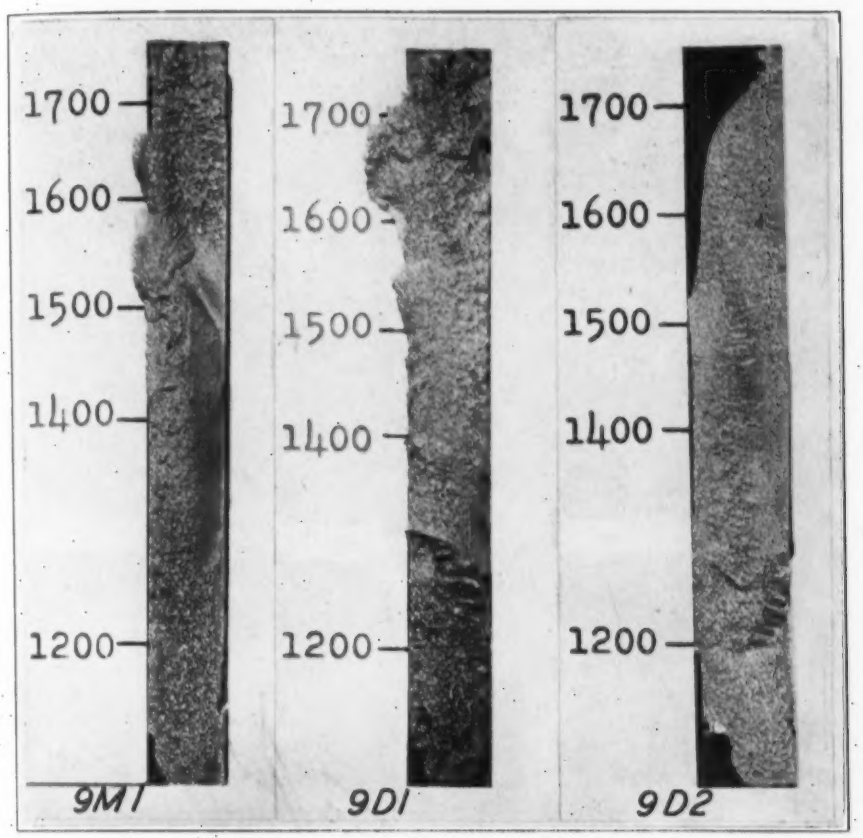

Fig. 3—Fracturegraphs of the Three Steels Shown in Fig. 1 Made on Specimen such as Shown in Fig. 2.
The figures denote temperature of the specimen in degrees Fahrenheit at time of quenching. The axis of the bar is toward the left; the surface is toward the right. Actual size.

to bring the three thermocouples up to predetermined temperatures, those used most regularly being 1200, 1450 and 1700 degrees Fahr. (650, 790 and 925 degrees Cent.). After some experience it has been found possible to heat to these temperatures in such a manner as to permit holding for about five minutes at temperature to insure thorough heating before the quench. The specimen is then removed from the furnace, grasped by tongs on the cold end and quenched in 15 per cent brine. It is then slotted to the axis on the grinder with a thin wheel as shown by the dotted line in Fig. 2, and carefully split open with a wedge. The three temperature points are marked on a card attached to the side of the specimen toward the axis. Additional temperatures desired above approximately 1400 degrees Fahr. (760 degrees Cent.) may be obtained by interpolation. Experience

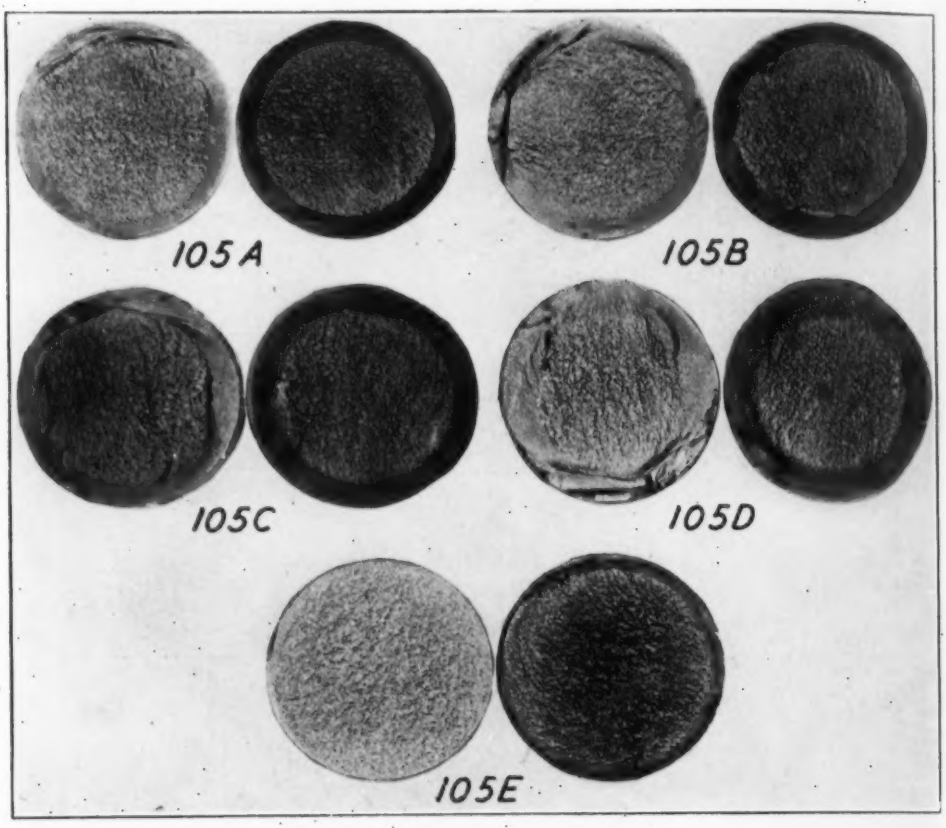

Fig. 4—Fracture Tests of Five Steels Containing Approximately 1.05 per cent Carbon Quenched in Brine at 1430 degrees Fahr. (775 degrees Cent.)
Fractures to the left are unetched. Those to the right are etched to bring out hardness penetration. Hardened in 1⅛ inch diameter section.

however has shown that it is not safe to interpolate below this temperature due apparently to absorption of heat in the portion of the bar just passing through the critical.

The resulting fracture gives a complete view of the quenching characteristics of the steel for approximately a ¾-inch diameter section. The test piece is convenient and compact, and after some experience in its use, will reveal almost at a glance many of the important hardening characteristics of the steel. Three such "fracture-graphs" made on steels 9M1, 9D1 and 9D2 are shown in Fig. 3, in which numbers to the left of each section denote temperatures in degrees Fahrenheit. It is interesting to note the sharp line about ¾ inch from the bottom of the piece separating the soft fracture below from the hard fracture above. The part above this line represents the portion of the bar which was under the critical point

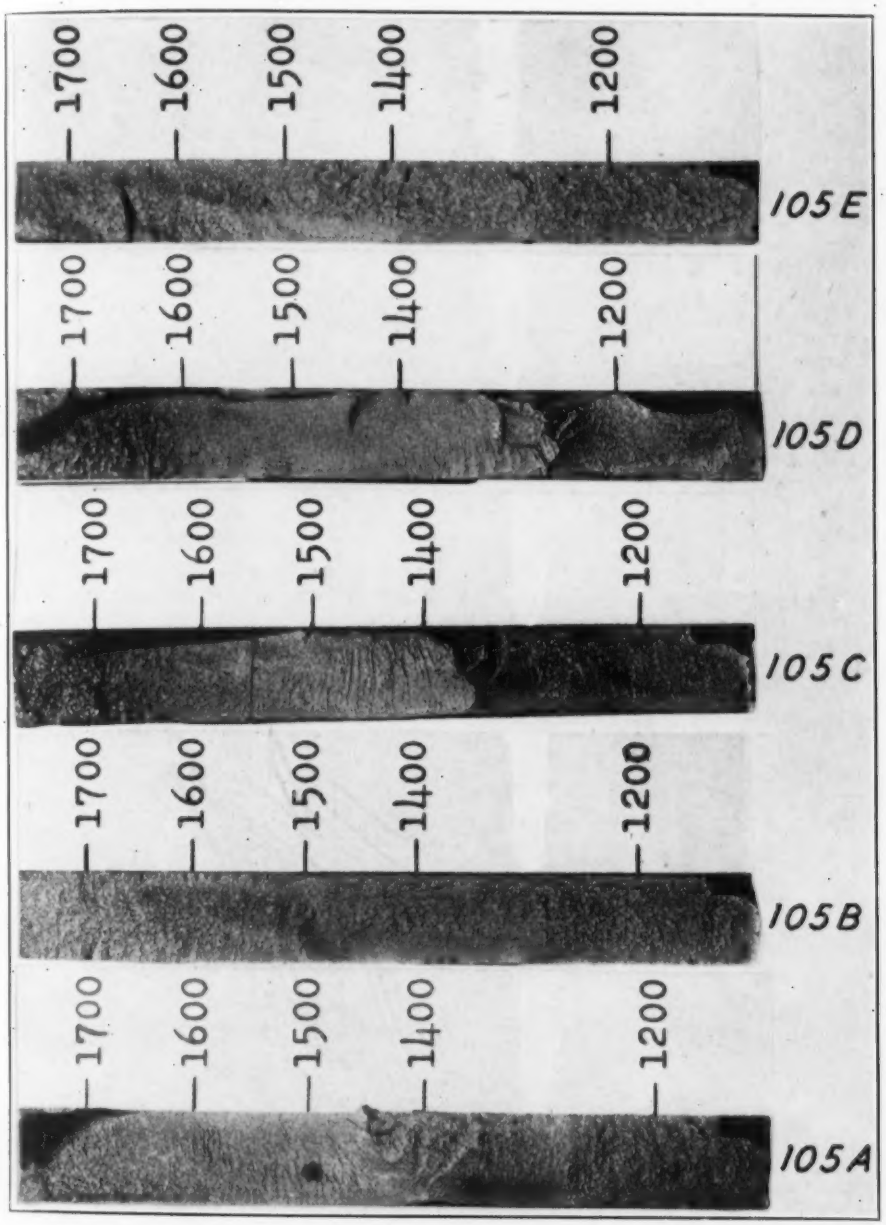

Fig. 5—Fracturegraphs of the Five Steels shown in Fig. 4. Photograph actual size.

at the time of quenching, while the latter shows the portion which was above the critical. The temperature figures in each case face the axis of the bar. A comparison of these fractures with those shown in Fig. 1 will be found interesting.

That the variations already illustrated are not confined to the

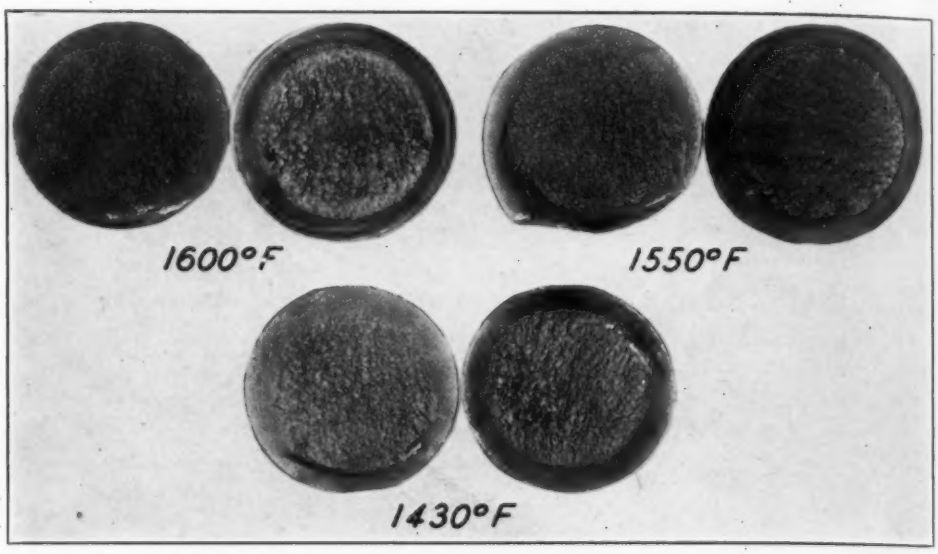

Fig. 6—Fractures of 95M Quenched in Brine at Three Temperatures.
Fractured surfaces on the right have been etched to show hardness penetration. Compare with Fig. 7. Actual size.

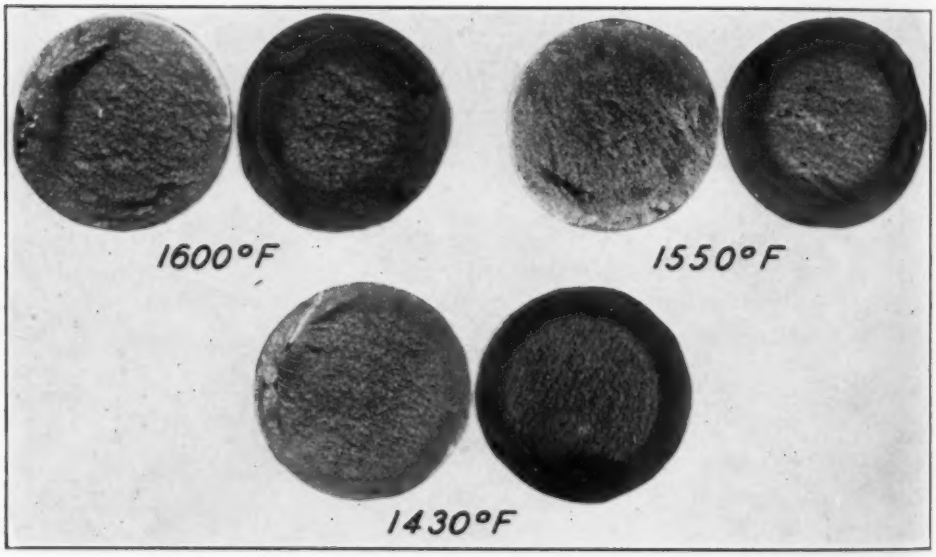

Fig. 7—Fractures of 95D Quenched in Brine at Three Temperatures.
Fractured surfaces on the right have been etched to show hardness penetration. Compare with Fig. 6. Actual size.

eutectoid range of carbon, but apply also to the hypereutectoid is shown in Figs. 4 and 5. Fig. 4 shows the natural and etched fractures of five steels quenched at 1430 degrees Fahr. (775 degrees Cent.) in a 1⅛-inch section, while Fig. 5 shows the corresponding fracturegraphs. The penetration on steel 105E is hardly visible, on

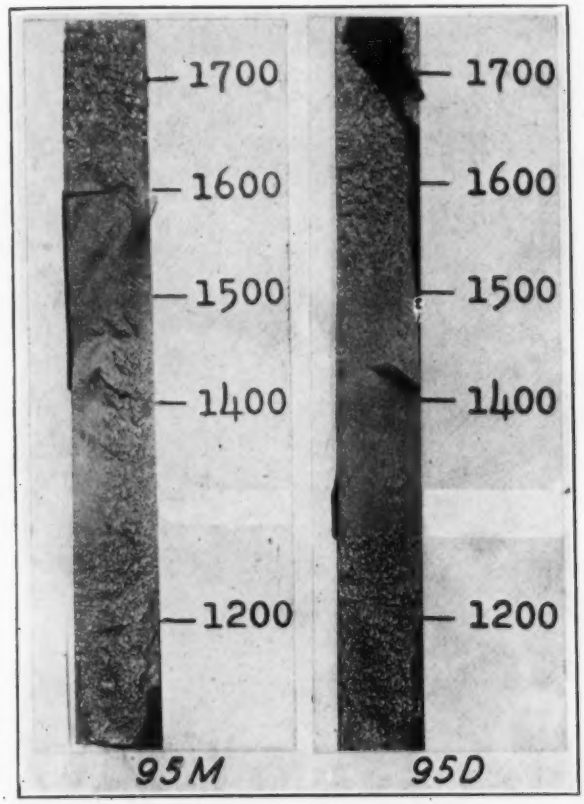

Fig. 8—Fracturegraphs of Steels Shown in Figs. 6 and 7. Axis of the bar is toward the right. Actual size.

the 1⅛-inch round bar amounting to only about $\frac{1}{32}$ inch. That on 105D is approximately $\frac{3}{16}$ inch as shown.

It is quite apparent from the above examples that the fracturegraphs show the quenching characteristics of the steel equally as well as the individual fractures, and will be used frequently in this paper to show penetration and quenching range of the several steels discussed.

Relation of Hardening Characteristics to Properties of Quenched Pieces

Referring again to the fractures shown in Figs. 1 and 4, one naturally wonders how steels of such widely different properties will behave when made into various tools and hardened by some uniform practice. Undoubtedly uniform results will not be obtained.

Steel 9D2 with a deep penetration will certainly behave differently than 9M1 with the shallower penetration when hardened

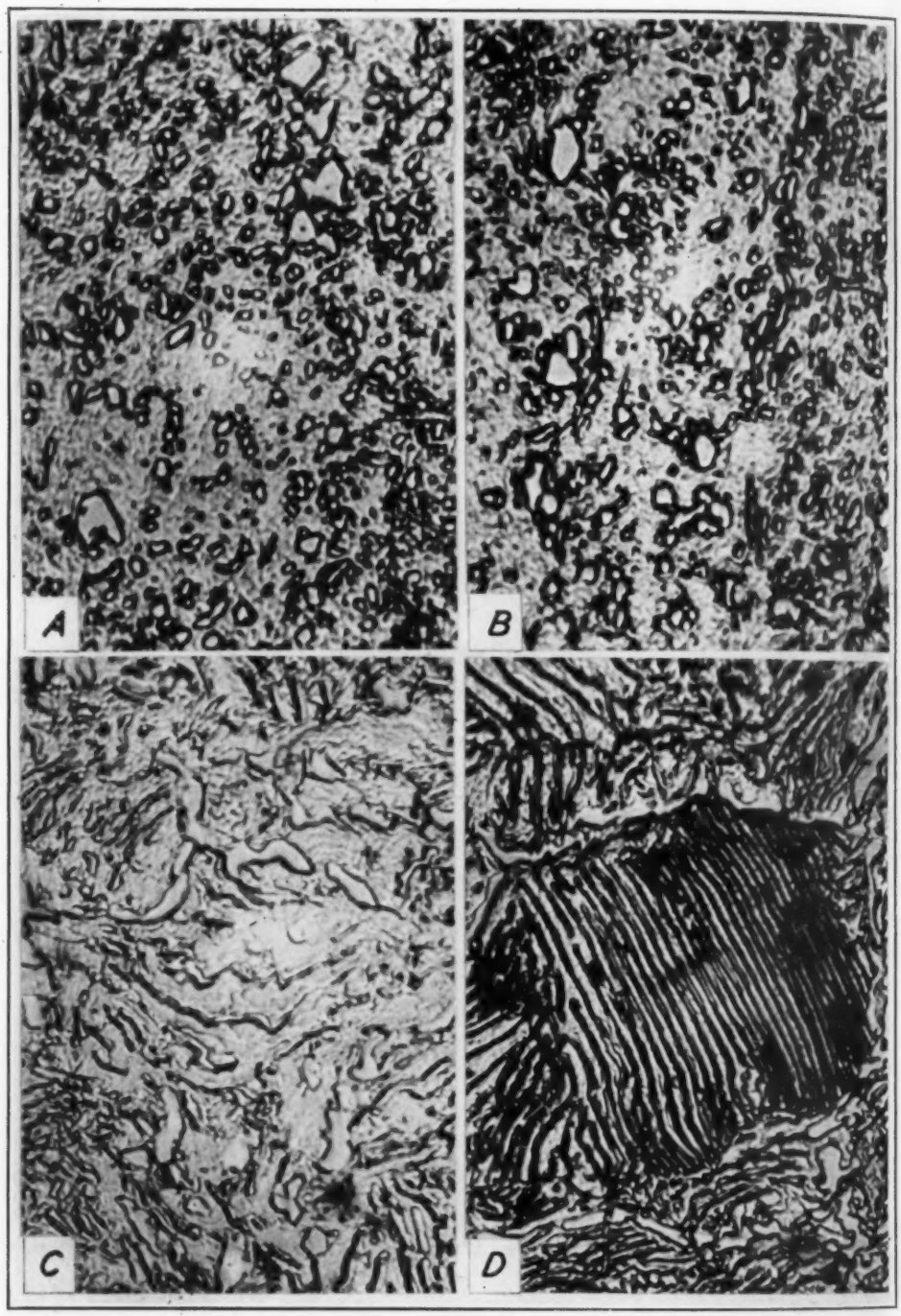

Fig. 9—(A) Annealed Structure of Steel 95M.
(B) Annealed Structure of Steel 95D.
(C) Steel 95M Carburized 6 hours at 1625 degrees Fahr. and Cooled in Pot. Section taken close to surface.
(D) Steel 95D Carburized 6 Hours at 1625 degrees Fahr. and Cooled in Pot. Section taken close to surface.
All photomicrographs × 1000. Etched 4 seconds in 2 per cent HNO_3 in alcohol.

at say 1430 degrees Fahr. (775 degrees Cent.), particularly if the tool contains sections of a critical size. Steels 9M1 and 9D1 would be expected to yield very dissimilar results when quenched at from 1550 to 1600 degrees Fahr. (845 to 870 degrees Cent.), the temperatures commonly employed in hardening such tools as header dies or striking dies. A clear idea of some of these differences can be obtained by quenching simple shapes under various conditions of size, temperature and quenching medium, and the following few paragraphs will be devoted to the study of two steels of similar composition but dissimilar hardening properties by this method.

Analysis and Structure. The two steels selected for this work contained approximately 0.95 per cent carbon and are designated 95M and 95D. Complete analyses are shown in Table I. Figs. 6 and 7 show fractures of these two steels after quenching at 1430, 1550 and 1600 degrees Fahr. (775, 845 and 870 degrees Cent.). In each case both halves of the fractured bar are shown, the half to the right having been etched to more clearly show the penetration of hardness. Fig. 8 shows corresponding fracturegraphs. Steel 95M will be recognized as a steel of average hardness penetration and wide quenching range. Steel 95D has a deep penetration with a narrow quenching range.

Fig. 9 shows the annealed structure of the bars from which the above pieces and all subsequent samples were cut. Brinell hardness of 95D was 187; that of 95M was 174. Fig. 9 shows also the structure close to the surface of specimens of both steels after carburizing six hours at 1625 degrees Fahr. (885 degrees Cent.), cooling slowly in the pot. Specimen 95M is seen to be quite abnormal to the McQuaid-Ehn test, while 95D is quite normal. This agrees closely with the result obtained by Epstein and Rawdon on deep and medium hardening steels[6], although it will be well to point out here that normality is associated usually with narrow quenching range rather than with deep penetration alone.

Figs. 10 and 11 show structures of the two steels after quenching at 1430, 1500, 1550 and 1600 degrees Fahr. (775, 815, 845 and 870 degrees Cent.). The difference in size of martensite needles in the two steels is quite marked, and readily explains the difference in quenching range evidenced by the fractures.

Formation of Soft Spots on Quenching. Blocks $1\frac{5}{16}$ inch square

[6] loc. cit.

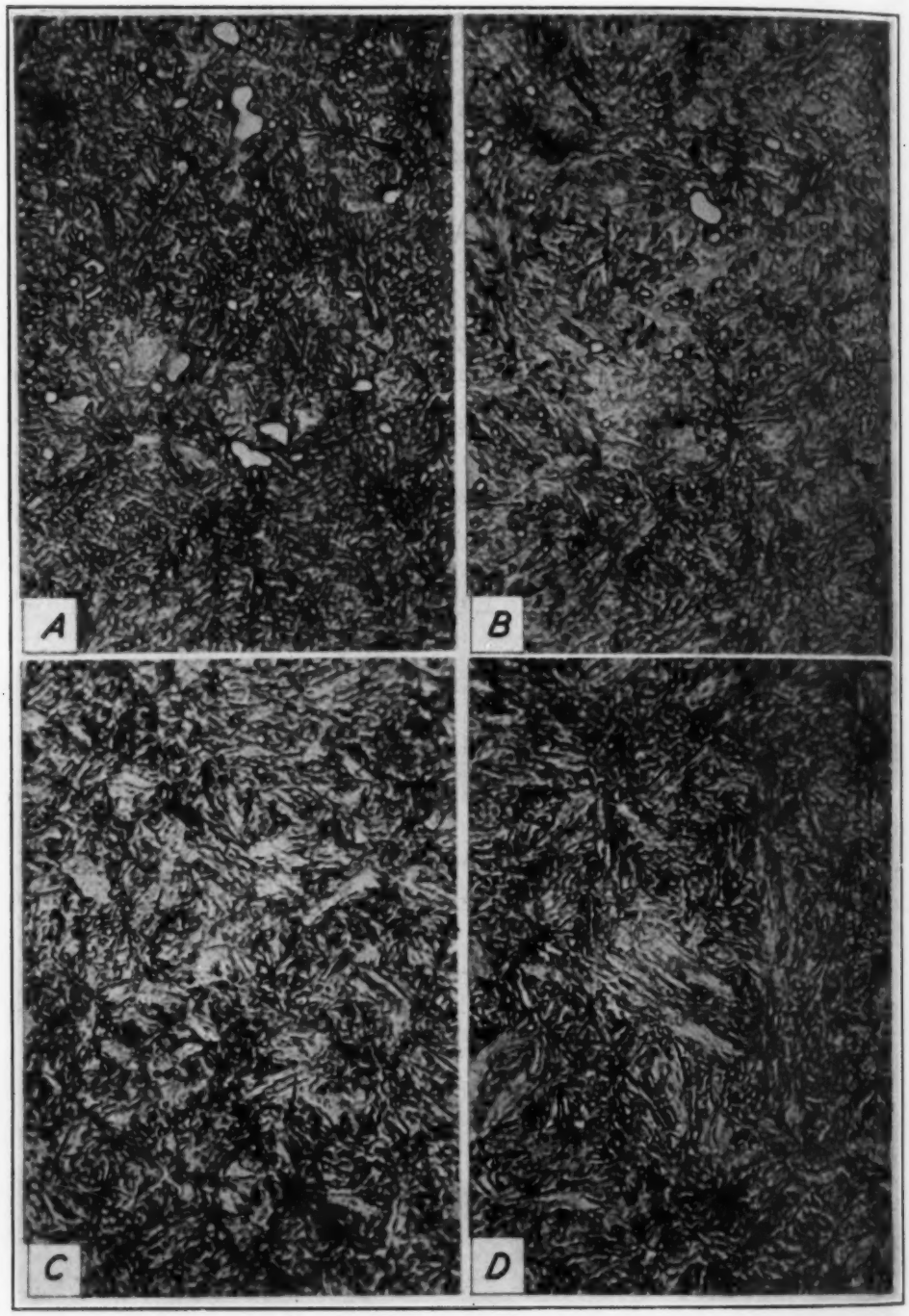

Fig. 10—Structure of Steel 95M after Hardening.
(A) Quenched in Brine at 1430 degrees Fahr.
(B) Quenched at 1500 degrees Fahr.
(C) Quenched at 1550 degrees Fahr.
(D) Quenched at 1600 degrees Fahr.
All photomicrographs at × 1000. Etched four seconds in 2 per cent HNO_3 in alcohol.

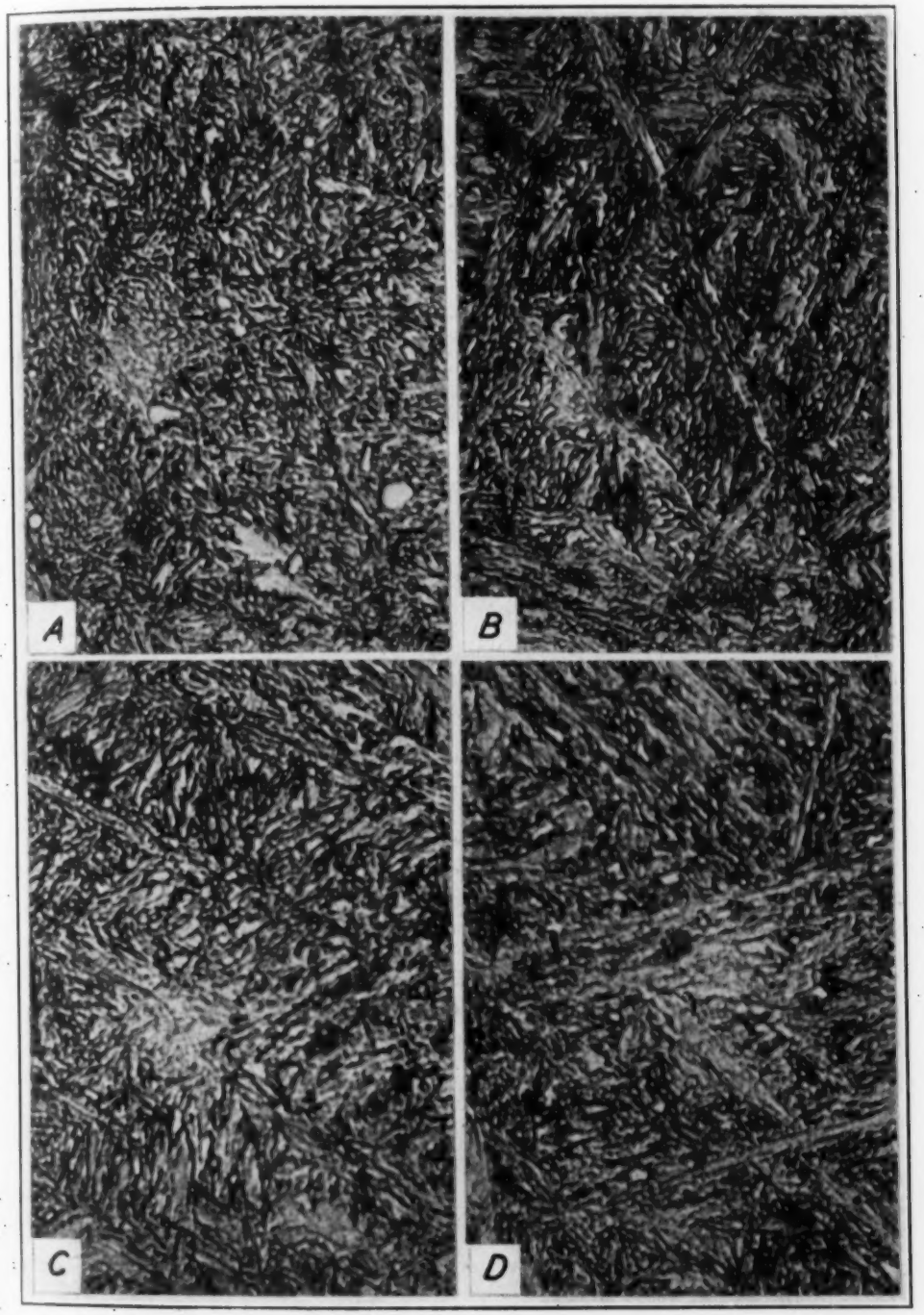

Fig. 11—Structure of Steel 95D after Hardening.
(A) Quenched in Brine at 1430 degrees Fahr.
(B) Quenched at 1500 degrees Fahr.
(C) Quenched at 1550 degrees Fahr.
(D) Quenched at 1600 degrees Fahr.
All photomicrographs at × 1000. Etched 4 seconds in 2 per cent HNO_3 in alcohol.

by 2 inches in length with chamfered edges 1/8 inch in width were cut from both steels and quenched from 1430 degrees Fahr (775 degrees Cent.) in 15 per cent brine in tap water with agitation, in tap water without agitation, and in water containing an abnormal amount of CO_2 introduced by bubbling gas through the bath for about an hour. The temperature of all quenching baths was held at 65 degrees Fahr. Pieces were grasped on the ends with tongs and quenched vertically, so that the four sides were in perfect contact with the quenching liquid.

All the brine-quenched pieces, as well as the pieces quenched in the agitated water hardened free from soft spots. The pieces from 95D quenched in still tap water hardened free from soft spots, while those from 95M quenched under the same conditions and at the same time, developed soft spots covering 5.8 per cent of the area of the four sides. All pieces of both steels quenched in water containing CO_2 developed soft spots. Those on 95M covered 61 per cent of the area of four sides, while those on 95D covered 39 per cent of this same area.

The appearance of the most interesting of the specimens is illustrated in Fig. 12. The patterns were developed by etching in 1:1 HCl at 160 degrees Fahr. (70 degrees Cent.) for five minutes. The white areas indicate soft spots while the dark areas show the hard martensite shell. It is interesting to note that regardless of the steel or the quenching bath used, the hardness of the soft spots was quite uniformly 42 to 44C Rockwell, while that of the hard shell was uniformly 65 to 67C.

These specimens bring out an interesting and important fundamental difference between the two steels, viz., their difference in critical quenching speed, 95D having a slower speed than 95M as evidenced by its deeper penetration and lesser susceptibility to soft spots on quenching. In other words the same characteristic which explains the greater freedom of 95D from soft spots also accounts for its deeper hardness penetration. Both are functions of critical quenching speed.

While 95D has a slower quenching speed than 95M, it does not by any means follow that it is the better steel. On the other hand there are very strong evidences that it is inferior to 95M, the latter having a finer, tougher hardened structure and a penetration which has been found satisfactory in actual practice. In any event it is

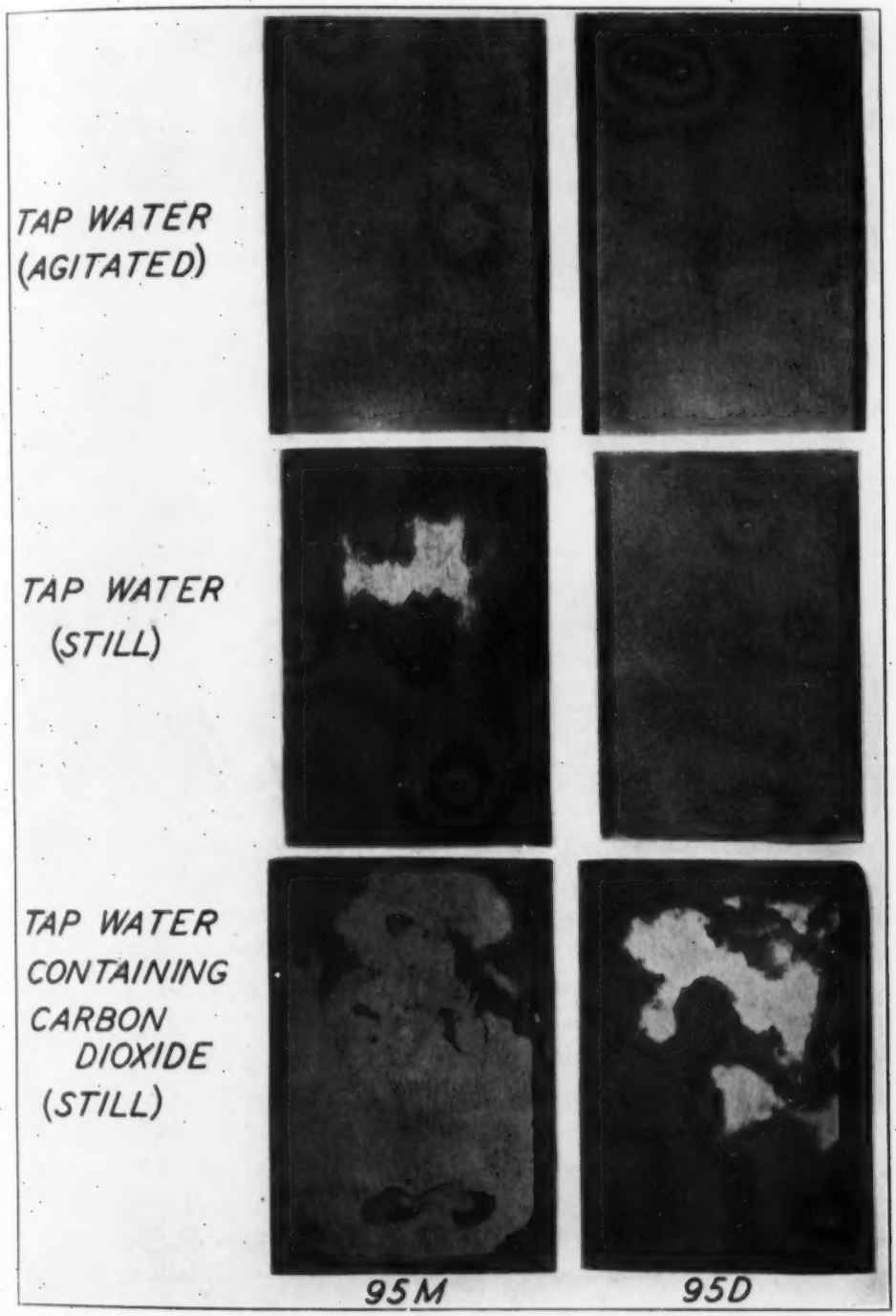

Fig. 12—Blocks from Steels 95M and 95D Quenched from 1430 degrees Fahr. in the Coolants Indicated.
Light areas indicate soft spots. Specimen etched in 1:1 HCl at 160 degrees Fahr. Photographs actual size.

not the purpose of the test to prove superiority of one steel over the other, but rather to demonstrate the necessity of uniformity in performance under a given method of quenching. It is quite apparent that either steel can be hardened free from soft spots if the proper method is employed. But it is also apparent that the quenching method which fits one steel will not necessarily fit the other, and that given a fixed method of quenching, the use of a mixture of both kinds of steel will inevitably lead to hardening room troubles.

Change of Size in Hardening. Changes in dimensions of tools during hardening are due to the fact that martensite, the hardened constituent in steel, occupies a greater volume per unit weight than do the softer constituents. Consequently if a cylinder of carbon tool steel of a section which will harden entirely through be quenched, it will be found of greater length after hardening than before. The behavior of a cylinder, however, which does not harden through is entirely different, and can best be understood by following the various changes which occur, beginning with the heating of the piece.

In heating for quenching, the entire piece expands in length to an extent which depends upon the thermal coefficient of expansion of the steel. Upon quenching, the first portion of the cylinder to contract is of course the outside shell, which upon contracting tends to pull with it the interior portions. These being still hot, and therefore plastic, are actually upset to a shorter length with consequent swelling in diameter. As the piece approaches atmospheric temperature in quenching, an expansion of the surface shell occurs due to the transformation of the outer layers to martensite. No such expansion, however, occurs in the core since softer constituents have already formed in this portion of the piece. After cooling to atmospheric temperature, therefore, the interior portions tend to be shorter than they were in the annealed piece, while the outside shell tends to be longer. Whether the piece as a whole finally shows an expansion after quenching, or a contraction, depends therefore on the depth of the martensite shell. If the shell is relatively thin compared with the entire section of the piece, the cylinder will show contraction. If it is relatively heavy it may show no change, or actual expansion. It is evident therefore that the properties of the steel affecting hardness penetration will have a marked effect on the change in dimension of the final tools. In order to illustrate this point, the following work was conducted upon cylindrical sections of steels 95D and 95M.

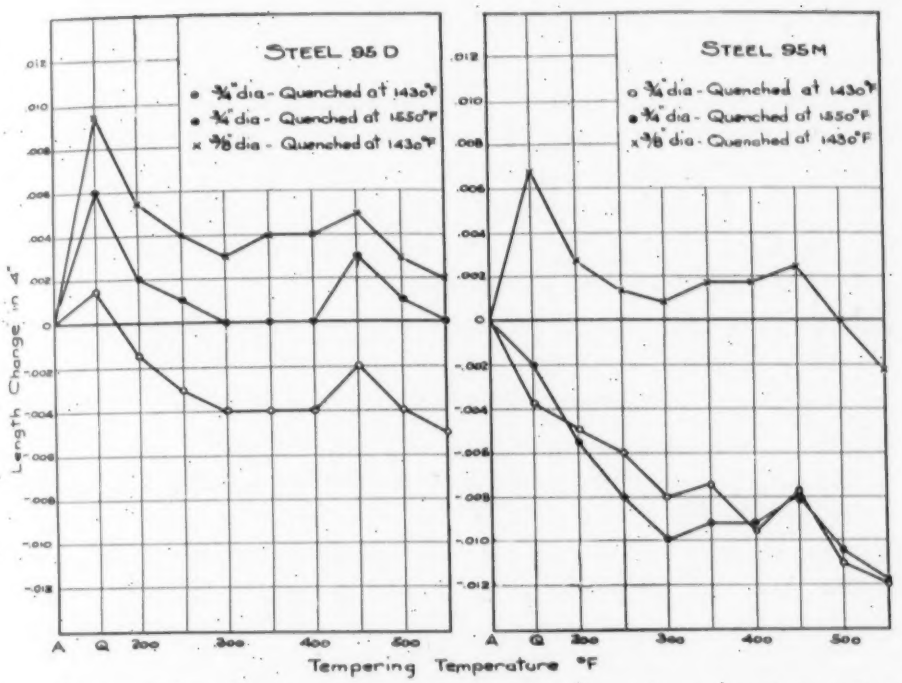
STEEL 95 D
• ¾" dia - Quenched at 1430°F
• ¾" dia - Quenched at 1550°F
× ⅜ dia - Quenched at 1430°F
STEEL 95M
o ¾" dia - Quenched at 1430°F
• ¾" dia - Quenched at 1550°F
× ⅜ dia - Quenched at 1430°F
Length Change in 4"
A Q 200 300 400 500
Tempering Temperature °F

Fig. 13—Curves Showing Length Change of Cylinders ¾ inch and ⅜ inch Diameter by 4 inches in Length after Quenching (Q) and after Tempering at Temperatures Shown.
Zero is taken as the length change in the annealed condition (A).

Several bars ¾ inch in diameter by 4 inches in length, and ⅜ inch in diameter by 4 inches in length, were accurately machined from both steels. Accurate measurements for length were made with a micrometer on all pieces in the annealed condition and recorded. The ⅜-inch diameter pieces were then carefully quenched in brine at 1430 degrees Fahr. (775 degrees Cent.) Some of the ¾-inch diameter pieces were quenched in brine at 1430 degrees Fahr. (775 degrees Cent.) and some at 1550 degrees Fahr. (845 degrees Cent.) All pieces were handled both in the furnace and in the quenching bath with thin wires to prevent any variations in the quench incident to handling with tongs.

After quenching, the pieces were again carefully measured for length and were then tempered in an oil bath for one hour 30 minutes at temperatures from 200 to 550 degrees Fahr. (95 to 290 degrees Cent.) in increments of 50 degrees, measurements being made after each successive temperature. Care was taken to allow each specimen to return to exactly the same temperature, viz., 70 degrees Fahr. before measuring, so as to avoid any error due to temperature effect. The results are shown graphically in Fig. 13.

These curves show plainly the marked difference in behavior between the two steels with regard to length change when quenched in a ¾-inch diameter section. Steel 95D having a characteristically deep penetration increases in length on quenching due to the preponderance of the expanding effect of the martensite over the shrinking effect of the softer constituents in the core, and it is noteworthy that the higher quenching temperature, resulting in this particular steel in considerably deeper penetration, is accompanied by the greater expansion. Steel 95M on the other hand shows a decrease in length on quenching due to the preponderance of softer constituents in the core over the relatively thin martensite shell, and the difference in quenching temperature here has little effect on account of the fact that it does not materially change the martensite penetration as has already been shown in Fig. 6. Thus we come to the conclusion that in sections over ¾ inch in diameter, steels of the nature of 95M are capable of being hardened at the higher temperatures not only without grain coarsening, but also without undue variation in size change. It should be pointed out in connection with the two steels that ¾-inch diameter is a rather critical section, and was selected for the purpose of showing in a striking manner the difference in behavior between two steels. It is evident that in the larger section, say 1½-inch diameter, steel 95D would also have decreased in length, although probably not as much as 95M.

The ⅜-inch section in both steels expanded on quenching for the obvious reason that both hardened entirely through. It will be observed, however, that consistent with the results of the larger sections, 95D showed a greater increase in length than did 95M.

Subsequent changes in length after tempering at increasing temperatures are seen to be fairly uniform, and within limits independent of the steel, the section or the quenching temperature, bringing out the interesting point that the length resulting from the quench determines the final length after tempering. Thus tempering cannot correct variations in dimension change, and it is necessary to have uniform changes in hardening in order to insure uniform dimensions in the finished tools.

The principles brought out above are very important in connection with the hardening of such tools as taps, in which even slight variations in lead change, in hardening, are extremely undesirable. They apply also to the hardening of sectional dies in which the change of size in hardening must be compensated for previous to

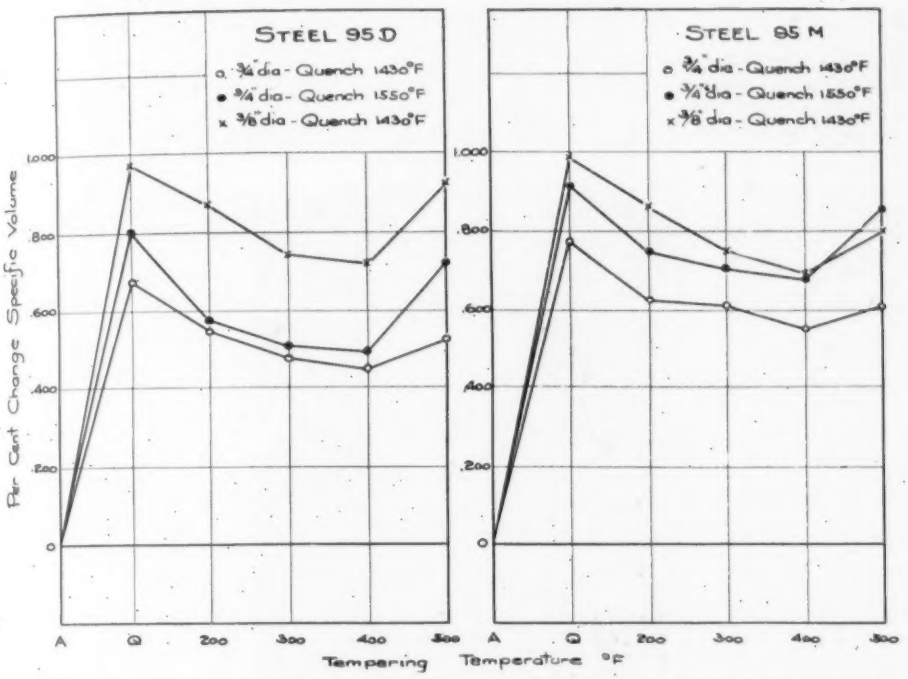

Fig. 14—Curves Showing Volume Change in Cylinders ¾ inch and ⅜ inch Diameter by ¾ Inch in Length after Quenching, (Q), and after Tempering at Temperatures Shown.

Zero is taken as the volume change in the annealed condition (A).

quenching so that the die will fit after hardening. These are only two of many possible examples which might be cited to show the necessity for uniform change of size in quenching.

The curves shown in Fig. 13 suggest another important point, viz., the possibility of a wide variation in the amount of internal strains set up in quenching. When the behavior of such strains is known, they may be compensated for both in the design of the tool and in hardening, and breakage in quenching or in service can be avoided. However, their behavior obviously cannot be predicted except in steel which behaves uniformly in quenching.

It is of further interest that the curves in Fig. 13 while in some cases based on specimens which did not harden through nevertheless check closely the results of other investigators on completely hardened specimens. Scott[7] for instance found the same characteristic increase in length at a tempering temperature of approximately 450 degrees Fahr. (230 degrees Cent.), attributing the original shrinkage on tempering to release of pressure, and the subsequent expan-

[7]Howard Scott, "Dimensional Changes Accompanying the Phenomena of Tempering and Aging Tool Steels." TRANSACTIONS, American Society for Steel Treating, Vol. 9, February, 1926, p. 277.

sion to the transformation of retained austenite to martensite following this release of pressure.

It should be borne in mind that length change in hardening must be considered as distinct from total volume change. The former is

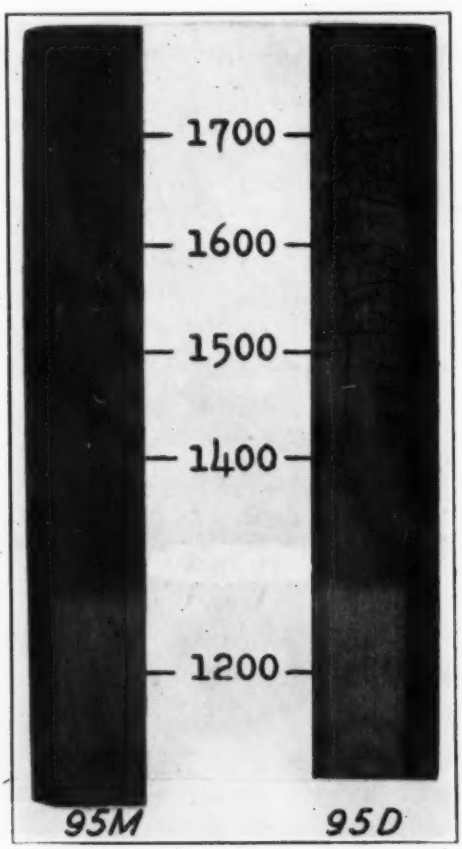

Fig. 15—Specimens Shown in Fig. 8—Surface-ground Under the Same Conditions, and Etched to Reveal Grinding Checks. Photographs actual size.

dependent upon the shape of the pieces being hardened. For instance while cylinders which harden with a relatively thin martensite shell will contract in length, they will expand in diameter. A cube of the same material will expand considerably across the middle of the faces, but only slightly on the edge. Volume changes, however, are independent of shape, and only dependent upon quenching temperature and character of material.

For purposes of comparison cylinders of the two steels ¾ inch in diameter by ¾ inch in length, and ⅜ inch in diameter by ¾ inch in length were carefully machined and weighed on an analytical

balance in air and in water, from which weights it was possible to calculate specific gravity. Some of the ¾-inch diameter sections were then quenched in brine at 1430 degrees Fahr. (775 degrees Cent.) and some at 1550 degrees Fahr. (845 degrees Cent.) The ⅜-inch diameter sections were quenched at 1430 degrees Fahr. (775 degrees Cent.) Specific gravity was determined in the quenched condition and again after tempering successively at 200, 300, 400 and 500 degrees Fahr. (95, 150, 205 and 930 degrees Cent.) Results are shown graphically in Fig. 14, in which per cent change in specific volume (reciprocal of specific gravity) is shown for the various conditions. This percentage is calculated upon the volume of the specimen in the annealed condition.

These curves show fairly uniform increases in volume in both steels in all cases. Both quenching temperature and size of section have their effect, but the order of these variations is slight. No explanation is given of why 95M showed a slightly greater volume increase than 95D. It is possible that the explanation may lie in some difference in the magnitude of internal strains resulting from the quench, and it is proposed to further investigate this possibility. As already stated these curves are shown for the purpose of preventing any possible confusion of volume change with length change in hardening.

Occurrence of Grinding Checks. The taper-quenched specimens of 95M and 95D shown in Fig. 8 were chucked fracture side down and end to end on a surface grinder, so that a flat area about ½ inch in width could be ground the entire length of both pieces, keeping the entire ground area within the hardened shell of the specimen. The travel of the table was parallel to the axis of the specimen. Grinding was done wet with a 36-grain 12-inch diameter wheel, having a feed of 0.040 inch, a cut of 0.005 inch at a speed of 1200 revolutions per minute.

The specimens were then etched for 5 minutes in 1:1 HCl at 160 degrees Fahr. (70 degrees Cent.) to reveal any possible grinding checks. The result is shown in Fig. 15. Steel 95M, with the wide quenching range, having a fine, silky fracture when quenched up to 1600 degrees Fahr. (870 degrees Cent.), shows no checks under 1650 degrees Fahr. (900 degrees Cent.) Steel 95D with a narrow quenching range is seen to check even in the section quenched at as low a temperature as 1400 degrees Fahr. (760 degrees Cent.), and checks badly in the regions quenched at higher temperatures. A

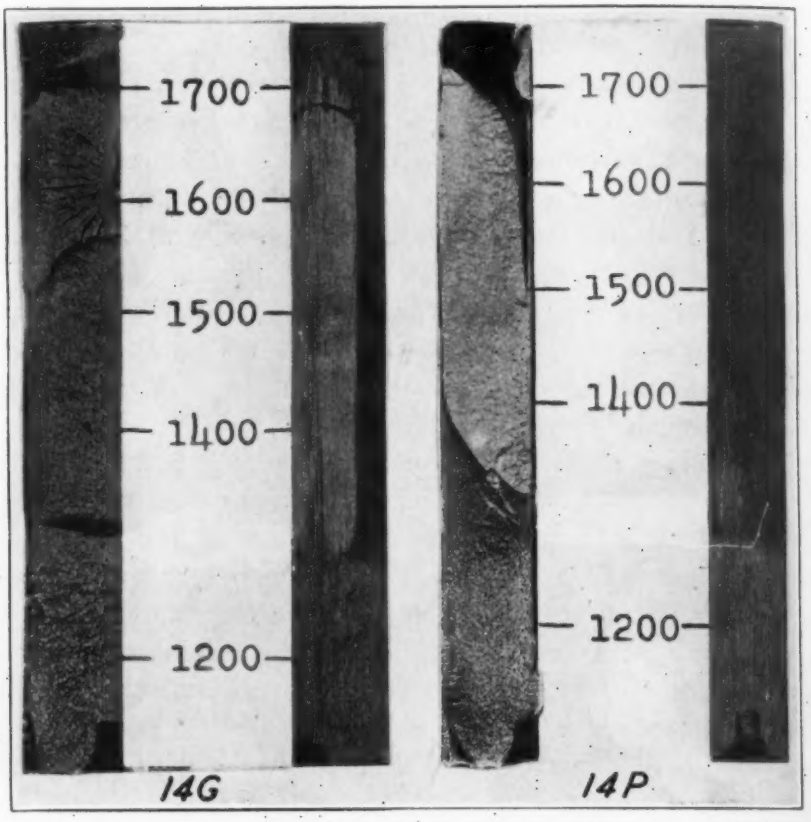

Fig. 16—Fracturegraphs of Two Steels Containing 1.40 per cent Carbon.
14G Gave Excellent Results in Granite Points.
14P Failed by Spalling of the Points. Fractures on the left. Etched sections on right. Photographs actual size.

direct comparison between Figs. 8 and 15 is interesting. It is evident that while proper tempering will probably prevent to a large extent the condition shown on specimen 95D, this condition is nevertheless undesirable, and one which might readily account for certain variations observed in the tendency of steels to check in grinding.

While in the foregoing paragraphs the study of the effect of variations in hardness penetration and quenching range is confined to pieces of simple section, the principles involved apply equally well to complicated sections such as are often encountered in actual tool practice. In fact for obvious reasons the effects shown on simple sections often are much more pronounced in the complicated sections.

It is indicated therefore that uniformity in hardness penetration and quenching range is necessary if uniformity in the final tools is to be obtained. It is also indicated, however, that steels of the charac-

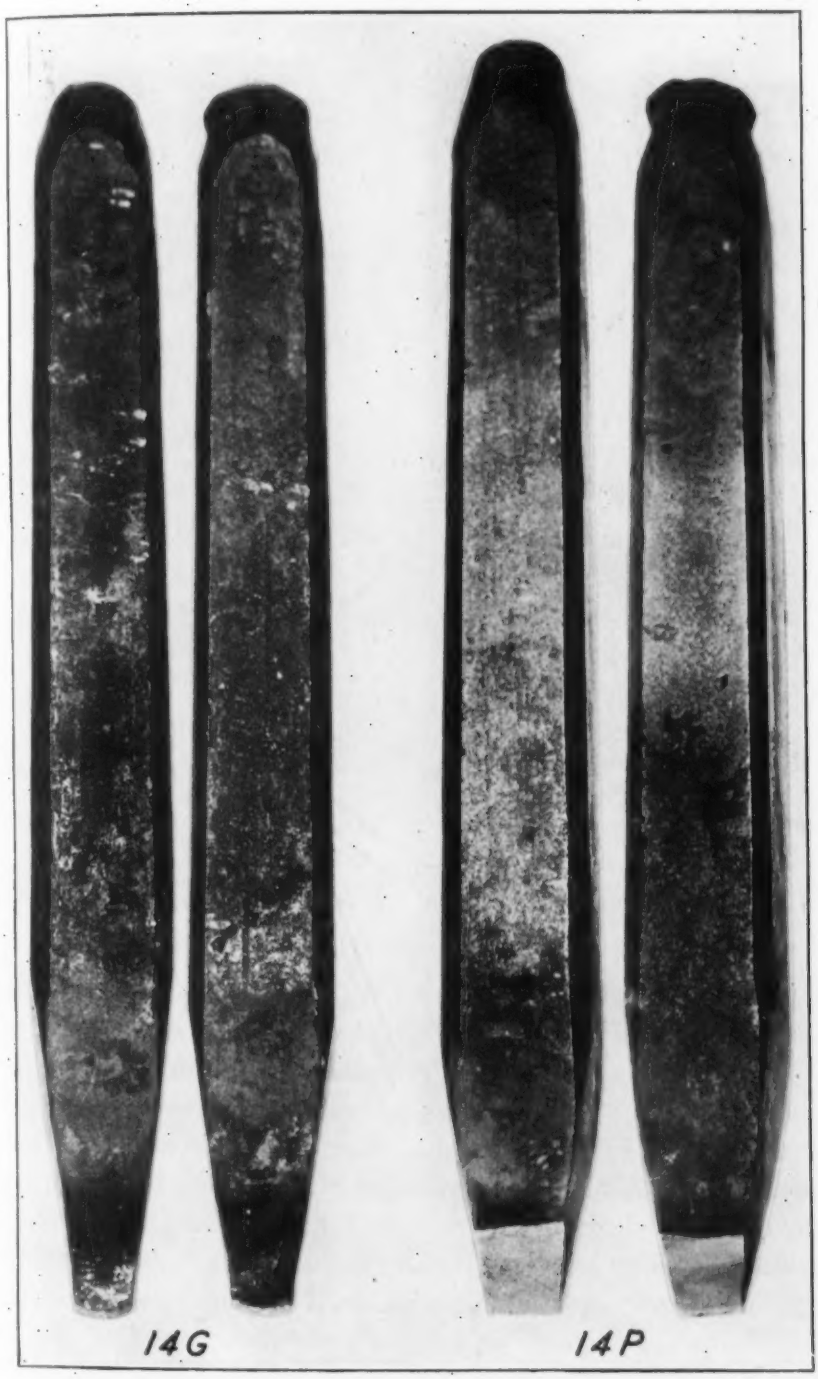

Fig. 17—Granite Tools from which Tests Shown in Fig. 16 were Taken. Note two types of fracture on spalled points of 14P. Photo actual size.

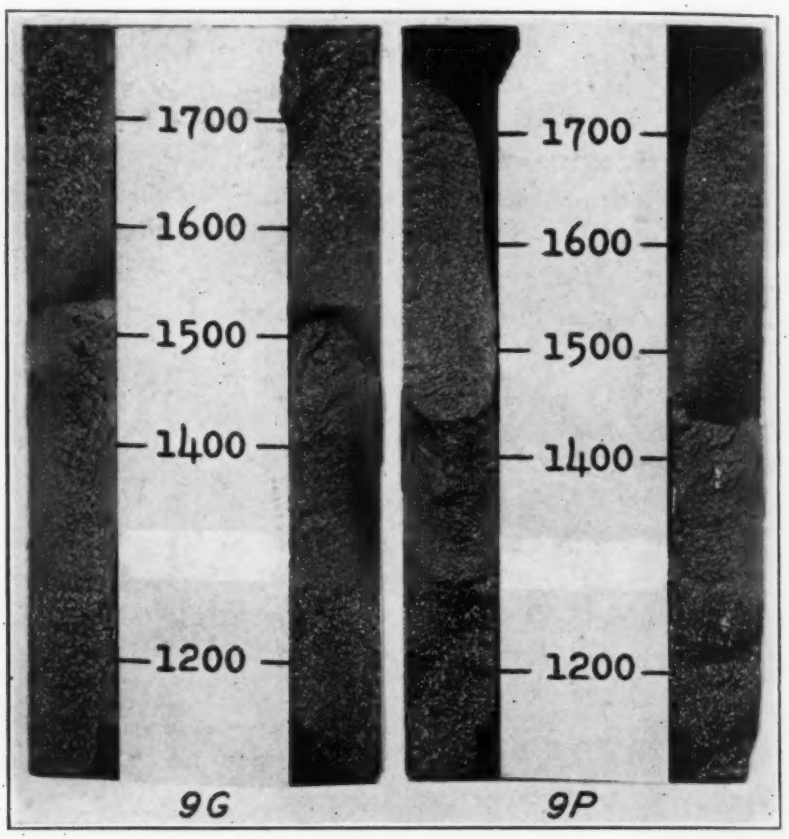

Fig. 18—Fracturegraphs of Two Steels Containing 0.90 per cent Carbon. 9G Gave Excellent Results in Swaging Dies.
9P Failed Due to Checking and Ultimate Fracture through the Impression. Photograph actual size.

ter of 95D should be avoided, and that for best average results steels of the character of 95M having good hardness penetration with a wide quenching range are most desirable. The characteristics possessed by the latter steel are such that accidental overheating will not impair the properties of the finished tool. While overheating should of course be avoided, there are still many tools being hardened under circumstances which do not permit of close temperature control, such as for instance those hardened from the forge fire. A heat-sensitive steel will usually give trouble under such circumstances, and often bring from the blacksmith or hardener the comment that the steel lacks body. Steels of the character of 95M, while undoubtedly tougher than the heat-sensitive steels when quenched at the high temperature, sometimes attained by accident but more often attained by design, such as in the hardening of certain dies, also appear

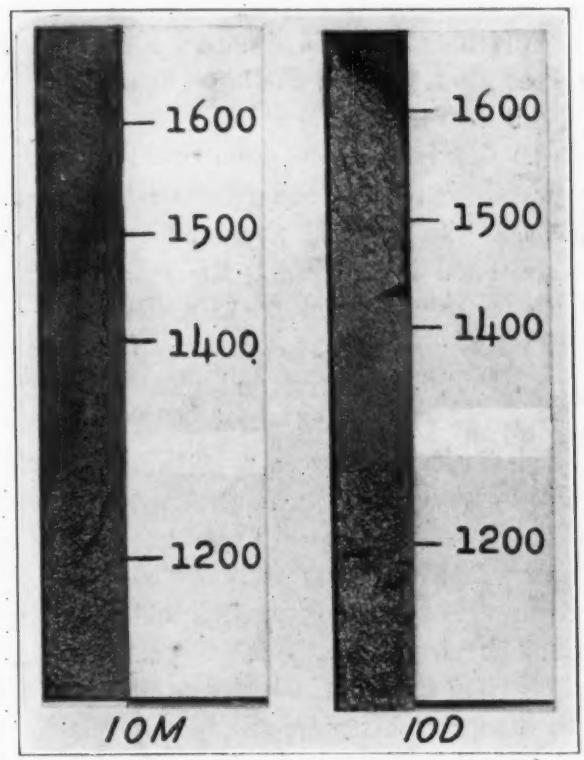

Fig. 19—Fracturegraphs of Two Steels Containing 1.00 per cent Carbon.
10M Gave Satisfactory Results in Cold Header Dies.
10D Either Cracked in Hardening or Failed after Making a Few Pieces. Photographs actual size.

tougher and more lasting in comparison with the other steels when quenched at the normal hardening temperatures.

To illustrate these points fully on actual tools would require a vast number of various types and sizes. A few simple cases, however, will suffice to show the general principles.

Relation of Hardening Characteristics to Actual Tool Service

Fig. 16 shows two steels of a type containing 1.40 per cent carbon commonly used in granite tools. The fracturegraph in each case is shown toward the left, and a corresponding section etched to bring out more clearly the penetration is shown to the right. It will be seen that 14G may be hardened from a temperature of 1650 degrees Fahr. (900 degrees Cent.) without appreciable increase in penetration. Steel 14P on the other hand hardens through in the

¾-inch section at about 1400 degrees Fahr. (760 degrees Cent.) The former retains a close-grained structure up to 1600 degrees Fahr. (870 degrees Cent.), while the latter is already coarse at 1450 degrees Fahr. (790 degrees Cent.)

Both steels in the form of ⅞-inch quarter octagon bars were submitted to a granite cutting concern for trial in hand chisels, with the request that both be handled throughout in the same manner. The treatment consisted in hardening the points from a forge fire from a dull red heat, and tempering to a light straw color. The chisels from 14G were reported the equal of any which they had ever used. Those from 14P, however, failed after only a few blows, the points spalling as shown in Fig. 17. It will be noted that of the two failed chisels, one has a coarse fracture, and the other a fine fracture, an interesting commentary on the necessity for a wide quenching range when tools are to be hardened from a forge.

It is of further interest in connection with these two steels that a sample of each was given to each of three blacksmiths, all in different shops, and all having had about 35 years' experience in the working of plain carbon steels. Each was asked to forge and harden the steels and to then give his opinion of their relative quality. All three reported that 14P had no "body," but that 14G had good body.

Fig. 18 shows fracturegraphs from two steels of identical composition as shown in Table I. Steel 9G is seen to have a wide quenching range with normal penetration and a fine silky grain. Steel 9P has a short quenching range, somewhat deeper penetration and a slightly dry fracture. Both were being used in a swaging die in which the impression consisted in a rather sharp notch cut through the die, and were assumed to be alike since the analysis was identical. However, early failures developed in some of the dies due to checking in the impression, followed by ultimate fracture. Tests cut from the bad dies showed the condition illustrated in 9P. Tests from the good dies all had the characteristics of 9G. The fracturegraph marked 9P was cut from a failed die which had to be scrapped after a run of three hours. The die from which 9G was cut had run 40 hours, after which time it was taken out for redressing.

Fig. 19 shows two steels of different hardening properties, 10M having a wide quenching range with normal penetration, 10D having a short quenching range, a temperature of 1430 degrees Fahr. (775 degrees Cent.) being sufficient to harden the specimen through in a ¾-inch diameter section with a dry coarse fracture. Both steels

were used in small solid header dies, operating on ¼-inch bolts. The dies were hardened by quenching in a flush at 1470 degrees Fahr. (800 degrees Cent.) and tempering at 425 degrees Fahr. (220 degrees Cent.) The dies made from 10D invariably cracked in hardening or failed by splitting after only a few bolts. Those from 10M made the normal number of pieces.

The examples given above involve steels of widely different characteristics resulting in widely different tool performance, and have been selected so that the results might be easily interpreted. However, differences of lower magnitudes also have their effect, perhaps not so noticeable, but bothersome nevertheless, when maximum tool service is the goal. The experience of several years has shown that practically all of these differences can be eliminated through the agencies of a uniform steel and a uniform hardening practice.

To summarize, then, the production of tool steel having uniform hardening characteristics is of advantage to designer, tool maker and hardener alike. For the designer it assures a definite and reliable penetration of hardness, making it possible to design his sections in such a manner as to reduce breakage and distortion in hardening to a minimum. For the tool maker it assures greater uniformity in dimension change, enabling him to make definite allowances for such changes. It also assures for him uniformity of performance in finish grinding. For the hardener it assures constant behavior in heat treatment, enabling him to more readily standardize his practice. With the introduction of more and more automatic equipment in the hardening room this has grown to be an important factor. It assures him of less breakage in hardening and less trouble in size change. Finally it will afford him a generous factor of safety on most of his operations.

In conclusion, it seems much within the scope of this society to foster cooperation between the steel manufacturer and the user, so that on the one hand a uniform tool steel will be supplied, and on the other a correct heat treatment will be applied. Many differences of opinion must needs arise between the maker and consumer, but it is believed that a common ground can always be reached by proper knowledge of the material. It is the hope in presenting this paper that it may help to bring many of the facts concerning carbon tool steel to light, not so much through the substance of the paper itself, perhaps, as through the discussion which it is hoped it will arouse, to

the end that this very useful steel may be applied with both greater satisfaction and greater economy.

The writer takes this opportunity to acknowledge the valuable assistance rendered by O. V. Greene, of the metallurgical staff, in preparing much of the experimental data used in the paper, and to express his appreciation of the help and cooperation extended him by the other members of the metallurgical department during the course of the work.

DISCUSSION

Written Discussion: By Frederick G. Sefing, Michigan State College, East Lansing, Mich.

It is indeed a notable occasion when more rigid specifications are recommended by the steel maker. The author is to be commended upon the simple and sure method of inspection which he presents. Various procedures are used by the consumers and makers for inspecting the hardenability of tool steels. It is important therefore that some specification be used which is acceptable to maker and user alike. Such a procedure of inspection must be simple and quick, yet dependable. The author's method fulfills these qualifications.

The writer is much interested to learn that steels of the character of 95M in the author's paper make desirable tools. Surely many of us would have rejected this steel because of its abnormal characteristics and consequently its tendency to show soft spots after hardening. Which proves again that "What is one man's food is another man's poison."

Too much emphasis cannot be laid upon the relation between length and volume changes with particular regard to the occasional decrease in length with corresponding increase in volume.

Written Discussion: By R. L. Dowdell, U. S. Bureau of Standards, Washington, D. C.

The author is to be congratulated on his exceptionally interesting and valuable paper.

It is apparent from Mr. Luerssen's results that tool steel users should keep various lots of steel separate and not mix the stock from various heats or mills. The fact that some tool steels have more "body" than others has long been known, but it is surprising to note that some industrial plants still mix steel stock of about the same grade and later expect to obtain uniform results on tools.

The author states that it is not within the scope of the paper to touch upon methods of manufacture which actually determines what the "body" of the steel will be. We would like to know more about the manner by which this "body" factor can be controlled in the mills, so we will greatly appreciate any information on this phase of the subject.

On page 164 the author implies that it is only partly true that hardness

penetration in quenched carbon steels is dependent entirely upon analysis, i. e., with regard to the elements usually determined. This is very likely a true statement and shows that some of the impurities not generally determined are responsible for the so-called "body" characteristics.

It appears that the various forms of abnormality are closely associated with both tool steels and low carbon steels for case hardening work.

Improved methods of chemical analysis are now being worked on by Louis Jordan and his collaborators of the chemical section, Bureau of Standards, whereby they expect to throw more light on this old question.

In working with steel castings we are finding that if we make final ladle additions of any of the different deoxidizers studied, that the mechanical properties of both the heat treated and the as-cast steels will be greatly affected, due apparently to submicroscopic nonmetallic inclusions formed during deoxidation.

I would like to ask the author whether or not his observations have indicated that fine nonmetallic particles display an important point in the so-called "body" of steel.

Written Discussion: By B. F. Shepherd, metallurgist, Ingersoll-Rand Company, Phillipsburg, N. J.

Mr. Luerssen is to be congratulated on placing this valuable information before the steel treaters' fraternity. It is hoped, as the author states, that this paper will draw forth much data which has been accumulated by the steel specialists. The published literature is rather meager in information of the character presented by Mr. Luerssen, which can readily be interpreted and applied to the solution of shop problems.

The stress placed upon the necessity of using ununiform tool steel to produce uniform product cannot be over-emphasized. Failures in hardening with ununiform hardening practice can many times be traced to ununiformity in the hardenability of the steel. Chemical analysis is only a guide to composition and not to hardenability. The product of one manufacturer may have a narrow range of hardenability, while that of another may have a wide range, whereas the average hardenability may be the same. Uniform product will be much harder to obtain with the wide range hardenability material.

The product of each manufacturer has characteristics directly related to the methods of manufacture used in that plant. Inasmuch as a brand name is used to identify a certain steel, we can readily see that a brand name identifies the particular characteristics imparted to the steel by the manufacturer. Brands of steel, therefore, are not interchangeable; even though the chemical analysis may be the same. This particular point must be impressed upon purchasing agents and others in order to reduce the tendency to purchase on basis of price and chemical analysis and to thoroughly protect the hardening department. Brand names indicate the quality characteristics of steel products.

The writer has presented a paper at this same session, which de-

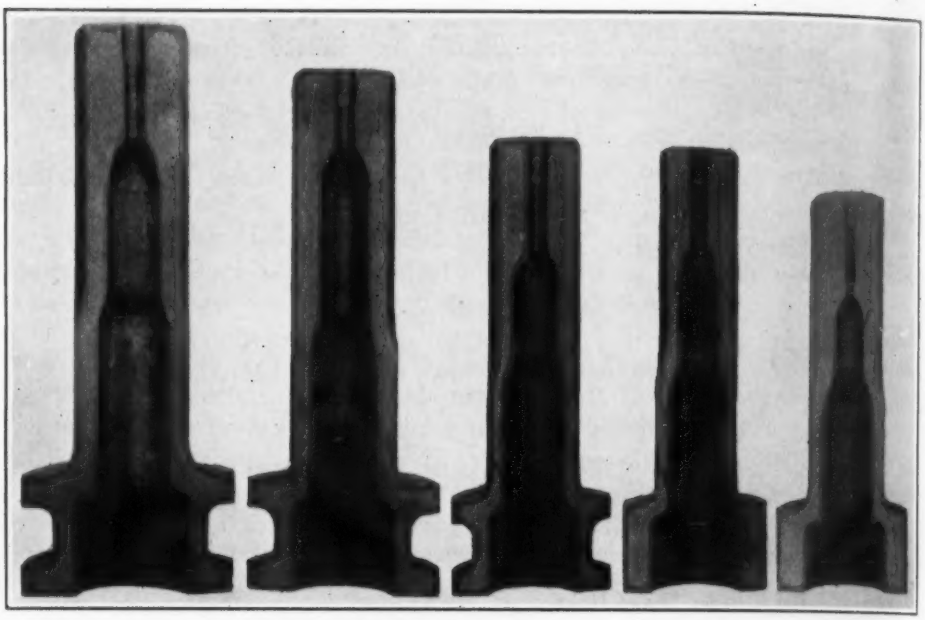

Fig. 1—Rock Drill Pistons Cut Apart and Etched.

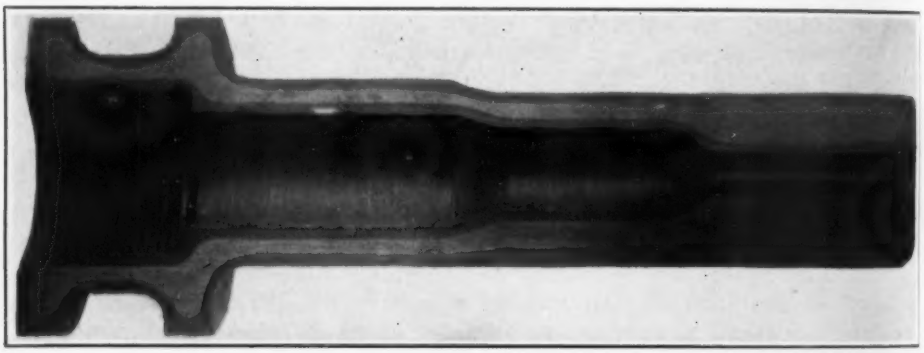

Fig. 2—Rock Drill Piston Showing Detail.

scribes a simple test method to determine differences in hardenability of steel. This test is being used regularly for inspection and material is being purchased specifying the desired hardenability. This control is necessary to maintain the highest quality of product. To illustrate what can be done with uniform steel and uniform hardening methods, two photographs are included. The first shows a number of rock drill pistons and the second, one of the pistons in greater detail. In both pictures the pistons have been split apart with a thin grinding wheel and etched to show distribution and regularity of the hardened zone which completely envelopes the part, inside and out. This zone is about $\frac{7}{64}$ inch thick.

The feature of temperature sensitivity mentioned by Mr. Luerssen has also been noticed in the writer's test. In general, the hardenability

number will be changed by an increase in quenching temperature of 75 degrees, while occasional heats will be received which take a higher hardenability classification with increase of only 20 degrees in quenching temperature. Many repeat tests have checked this feature and definitely identify it as temperature sensitivity.

The field of research in carbon steels has by no means been exhausted as suggested by Mr. Luerssen. There are applications in the writer's experience where straight carbon tool steels have shown superiority to any type of alloy steel tried. These straight carbon tool steels, of course, were subjected to highly specialized inspection.

Written Discussion: By J. J. Desmond, North East Electric Company, Rochester, N. Y.

The author is to be congratulated for the excellent paper in which he points out the different properties of carbon tool steels of the same approximate chemical and microstructure with their resultant effects upon tools made therefrom. It is important to note that carbon tool steels of similar composition act so differently upon quenching and the data presented will go a long way toward explaining some of the variations observed in the hardening of tools.

The writer cannot agree with the author that tools made of carbon steel are simple to harden. It is his opinion that many tools made of carbon steel require all the skill of the hardener in order to keep them in one piece. This is especially true unless every attention is given to the design of the tool. As is well known, a misplaced screw hole, the absence of a fillet, etc., is often responsible for the loss of a tool. It is the danger of tools made from carbon tool steel to crack while hardening that has driven many to adopt the more simple oil hardening steels.

The modified Metcalf test, as described in the paper, undoubtedly provides a great deal of information concerning properties of the steel. It appears to be a rather difficult test to apply to the general run of sizes of steel purchased. The hardness and fracture tests at two or more temperatures and possibly the McQuaid-Ehn test in conjunction with others would it seems be more suitable for the general run of material.

What value, if any, has the author found the hot acid etch test for making quality tool steel?

It might be well for the author to state the maximum and minimum depth of case obtainable and the limits within which any particular depth may be controlled. This has reference to carbon tool steel as made by the controlled melting process as referred to in the paper.

Oral Discussion

A. H. D'ARCAMBAL:[1] Being connected with a concern that hardens many thousands of tools daily made of plain carbon tool steel we are very much interested in the information given out this afternoon in the two papers just presented. We would be interested in learning if this condition, that is the

[1]Director of A. S. S. T. Pratt and Whitney Co., Hartford, Conn.

depth of hardening, can be easily controlled. It is interesting to note that tool steel containing as little as 0.50 per cent chromium will show a change of hardening in the opposite direction to tool steel containing no chromium; for example, a 1-inch tap made of plain carbon tool steel will show a change in one direction whereas the same design of tap made from steel containing 0.50 per cent chromium will show a lead change in exactly the opposite direction.

W. N. VAN TASSEL:[2] This paper is important to all users of tool steel. My comments are from the point of view of a tool steel user with experience in hardening and fracturing test pieces as a part of routine inspection, but not with such exact and extensive tests as Mr. Luerssen's. One point has not been sufficiently emphasized, I believe;—that the steels dealt with in this paper are extremes. They are not typical of the run of tool steel made by reputable mills. In the variations that I have found in hardening steels made by various mills there has usually been more variation in steel coming from different mills than there has been in the products from any one mill. Different brands of carbon tool steel made by the same mill usually have about the same hardening characteristics, although there may be marked differences in other respects. The effect of a Swedish base has been mentioned. This does not seem to be a predominating factor. I have not seen any uniform difference in the hardening characteristics of steel from mills which lay a great deal of emphasis on the character of the base used, as compared with steel from mills which do not. There is, however, a great deal of difference between various mills as to the hardening characteristics of the steel which they set out to make as best for general-purpose use. We have seen that difference of opinion here this afternoon between one user and the views expressed by Mr. Luerssen. Except for a few exceptional jobs (notably some types of cold heading) the steel which consistently gives a fine grain gives better results in my experience than the steel which coarsens up rapidly as the hardening heat is raised.

Author's Closure

I wish to thank Mr. Sefing for his helpful remarks. It has been quite general experience with us that abnormal steels are more desirable for tools than the so-called normal steels. It should be pointed out here, however, that we must be very careful of our terminology in discussing normal and abnormal tool steels, since the terms normal and abnormal were developed originally in connection with carburizing steels. Our conception of a normal tool steel is one which when heated to 1625 to 1650 degrees Fahr. and cooled slowly, will develop large grains of pearlite, while an abnormal steel is one which, under the same conditions, will develop small grains of pearlite. There are some carburizing steels which are termed fine-grained normal, so that in carrying this terminology over to carbon tool steel it is necessary to use some caution.

Mr. Dowdell has brought up the question of how "body" in tool steel can be controlled. It seems to me that the simplest way to answer this question would be to say that if you wish to make a steel having a wide quenching range, make it abnormal to the McQuaid-Ehn test. Tool steel which when

[2]Member of A. S. S. T. Scovill Mfg. Co., Waterbury, Conn.

cooled slowly from a temperature of 1600 degrees Fahr. shows under the microscope a structure consisting of large grains of well developed pearlite, will if quenched from this temperature have a coarse brittle structure. Heating to 1600 degrees Fahr. develops in this character of steel a very coarse austenitic structure. Upon slow cooling the coarse austenitic grains transform to coarse grains of well developed pearlite, characteristic of normal steel, while upon quenching they transform to coarse grains of martensite producing a coarse, dry and brittle fracture when the specimen is broken. Similarly a steel which when cooled slowly from a high temperature shows under the microscope a structure consisting in fine grains of pearlite, usually divorced, will if quenched from the same temperature have a tough fine-grained martensitic structure. Consequently steel with good body may be considered as synonymous with abnormal steel.

In answer to Mr. Dowdell's question regarding the influence of fine non-metallic particles on "body," we have a very strong conviction that such particles play an important part in inhibiting grain growth. It is evident, however, that these must be of colloidal dimensions to produce such effects. This explanation does not differ greatly from Mr. Ehn's original theory to account for abnormality.

Turning now to the factors controlling normality, steel manufacturers today are reasonably familiar with the methods employed in making steels of various degrees of normality. It is a well known fact that it is difficult to make abnormal steel having a high manganese content, say over 0.45 per cent. It is also well known that additions of aluminum promote abnormality. Increases in silicon content over 0.45 per cent will deepen penetration, while a silicon content of less than 0.10 per cent, other things remaining constant, will result in very shallow penetration. However, what should interest us most is to produce uniform steel from heat to heat and this is accomplished by careful attention to a great many details of practice. The character of scrap charge, methods of working the slag and method of deoxidizing all are important. Working and tapping temperatures are also of importance and must be controlled accurately by pyrometers. Having established a well defined practice, the details must be accurately repeated from heat to heat if uniform steel is to be produced. The effect of variations in practice is brought out nicely by Mr. Van Tassel, who points out that variations encountered between different makes are greater than variations encountered in the steel of any one maker, showing how strongly the practice peculiar to one manufacturer is reflected in the characteristics of the finished product.

In answer to Mr. d'Arcambal's question, hardness penetration and quenching range can be controlled at will by regulating and controlling the factors discussed above.

Mr. Van Tassel points out that the examples shown in the paper represent extreme conditions. This fact is commented on in the paper and it is explained that widely different steels were selected in order first to illustrate the variations possible, and second to bring out more forcibly the effect such variations have on the finished tools. It is reasonable to expect that smaller variations will have similar effects, less marked perhaps than those illustrated, but

important when measured in terms of tool service on long runs. Mr. Van Tassel's remarks on the effect of Swedish base on the performance of tool steel are very interesting.

Mr. Shepherd's remarks are greatly appreciated, since they represent the result of similar investigations from the viewpoint of a user of steels made by various mills. Mr. Shepherd's conclusion that chemical analysis is only a guide to composition and not to hardenability confirms the writer's experience. The photographs illustrating his method of showing hardness penetration in pistons are extremely interesting.

The writer agrees with Mr. Desmond that many tools made from carbon steel require considerable skill on the part of the hardener in their treatment, and that this fact has driven many to adopt alloy steels. This condition was remarked upon in the paper, and it was pointed out that variations in carbon steel probably partly account for the drift to alloy steels. The statement that carbon tool steels are simple to harden should possibly be modified to read that they are simple to harden if properly made.

Hardness and fracture tests at two or more temperatures in lieu of the modified Metcalf test have been found very satisfactory. In the writer's experience two pieces 1 inch in diameter, one quenched in brine at 1450 degrees Fahr. and one at 1550 degrees Fahr. and fractured, give a great deal of valuable information.

The hot acid etch test has been found of great assistance in making quality tool steel. It has been found of most value when applied to the billet stock before rolling, affording not only a means of preventing defective material getting into production, but also a reasonably prompt check on melting practice. The deep etch test is of course used to detect only material defects, and cannot be employed in determining any of the inherent hardening properties discussed in the paper.

In answer to Mr. Desmond's questions regarding penetration, the widest limits of penetration thus far encountered by the writer on steels of the same composition are 1/32 and 3/16 inch, both steels being quenched in brine in a 1-inch section from a temperature of 1450 degrees Fahr. Penetration may be controlled commercially within 1/32-inch as measured on a 1-inch diameter section.

In conclusion I wish to thank all those who have contributed to the discussion of this paper. The very excellent discussions certainly seem to indicate that the question of uniformity in carbon tool steel is a very live one, and a topic worthy of further study. In this respect it is hoped that the paper may prove to be a stepping-stone toward more exact knowledge of this steel and more complete understanding between user and manufacturer.

AUSTENITE AND ITS DECOMPOSITION

By Dr. Albert Sauveur

Abstract

An attempt is made to show that the mechanism of the transformation of austenite remains the same whether the cooling be slow, resulting in a pearlitic structure, or rapid, producing a so-called "martensitic" structure. The transformation is brought about by an alpha phase locating itself at the grain boundaries and along some of the crystallographic planes of the mother austenite, giving rise to the Widmanstatten type of structure on slow cooling and to the martensitic type on rapid cooling. Martensite is an aggregate of supercooled austenite and of a supersaturated solution of carbon in alpha iron—the latter containing also finely divided cementite particles.

In the author's opinion this alpha constituent answers our conceptions of troostite. The recent theories advanced to explain the hardening of steel by quenching, namely the grain-size, distortion and dispersion theories are briefly outlined and attention is called to significant differences between the hardening of steel after rapid cooling and the hardening of some nonferrous alloys after quenching followed by aging.

THE presentation of a paper on a subject which has been so frequently described and discussed demands some justification. It will be found in the consideration that differences of opinion still exist in regard to the mechanism and significance of the fundamentally important phenomenon of the decomposition of austenite.

It will be well at the outset to consider how deeply our material civilization would be affected, if austenite failed to transform in cooling through its thermal critical range. All steels, regardless of their carbon contents, would remain austenitic. They would be very difficult to machine and would have a limited field of usefulness. The heat treatment of steel would have little, if any, significance, and the American Society for Steel Treating would never have come into existence. Since all grades of iron and steel would

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remain nonmagnetic, electric motors and dynamos could not be constructed and the world would be practically deprived of electricity. It is a wise and kind providence which decreed that austenite should transform into a form very much more useful, and the reflection that this transformation consists in a slight shifting of the positions occupied by the iron atoms increases our wonder.

Austenite is generally regarded as a solid solution of carbon or of the carbide Fe_3C in gamma iron. Since its decomposition is brought about by the allotropic transformation of the solvent, it will be profitable to consider first the transformation of iron as free from carbon and other impurities as it can be obtained.

When pure iron solidifies at some 1530 degrees Cent. (2785 degrees Fahr.) it forms so-called body-centered cubic crystals; that is crystals built up of cubes with one atom at each corner and one atom in the center (Fig. 1). It is then known as delta iron, and is probably like alpha iron between A_3 and A_2, nonmagnetic; that is, paramagnetic. At some 1400 degrees Cent. (2550 degrees Fahr.), called the A_4 point, the atomic arrangement of the crystals quite abruptly changes and new crystals are formed consisting of cubes with one atom at each corner and one atom in the center of each of the six faces of the cube (Fig. 1). This arrangement is known as face-centered cubic and the iron now is called gamma iron. It is nonmagnetic. Observing that the face-centered arrangement is more closely packed than the body-centered, we infer that the transformation of delta into gamma iron must be accompanied by a contraction. According to some calculations made by Dr. R. H. Aborn (See Appendix), the relative volumes occupied by one atom of delta and of gamma iron are, respectively, 12.58 and 12.46 and the densities, respectively, 7.31 and 7.37, which confirms the expectations that the delta-gamma transformation is accompanied by a contraction. At some 900 degrees Cent. (1650 degrees Fahr.) occurs the critical point A_3 when the crystals again become body-centered, the iron, however, remaining nonmagnetic. It is now called alpha iron and was at one time called beta iron. It is well known that the gamma-alpha transformation is accompanied by a marked expansion which agrees with the observation that it now passes from a more closely packed (the face-centered type of space lattice) to a less closely packed (the body-centered type) arrangement. Dr. Aborn finds that the relative volume occupied by one atom of gamma iron at 900 degrees Cent. (1650 degrees Fahr.) is 11.66, while that of one atom of alpha

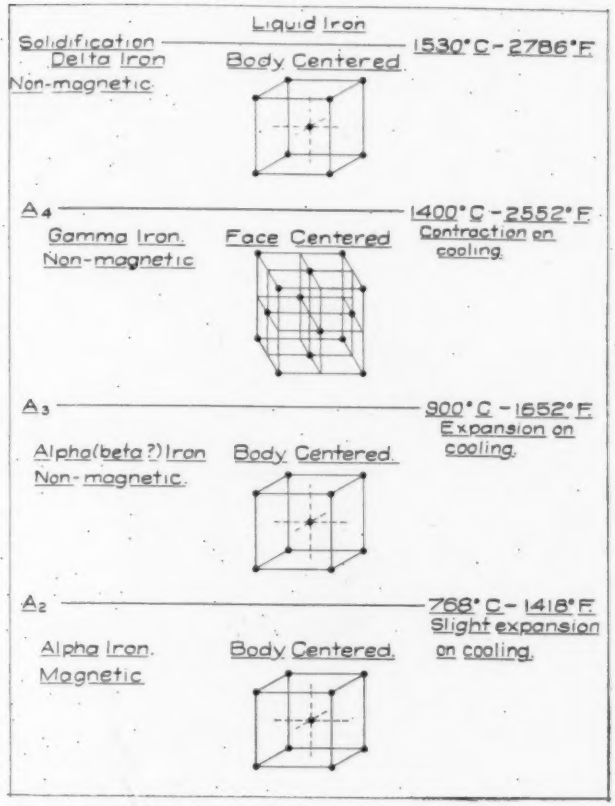

Fig. 1—Diagram Showing the Atomic Arrangement of Iron at Various Temperatures.

(beta) iron at 900 degrees Cent. is 12.26, their respective densities being 7.86 and 7.49.

At some 768 degrees Cent. (1415 degrees Fahr.)—the A_2 point—no change of space lattice occurs, the iron remaining body-centered, but it now becomes strongly ferromagnetic.

It is well to consider that we have apparently two kinds of body-centered iron, one which is paramagnetic and one which is ferromagnetic. The former is found at all temperatures exceeding some 1400 degrees Cent. (2550 degrees Fahr.) and also between 768 and 900 degrees Cent. (1415 degrees and 1650 degrees Fahr.), while the latter at all temperatures below 768 degrees Cent. (1415 degrees Fahr.) The argument that the loss of magnetism on heating starts at a temperature considerably below 768 degrees Cent. (1415 degrees Fahr.) while the recovery on cooling starts at a temperature considerably higher than 768 degrees Cent. (1415 degrees Fahr.) does not satisfactorily explain the marked evolution or absorption of heat oc-

curing at the A_2 point accompanied by a sudden change in its magnetic properties. One is led to believe that such a change in internal energy must be accompanied by some change in the space lattice, although undiscoverable by our present methods of X-ray analysis, or possibly by some change in the structure of the atoms themselves. There appears to be no good reason why the paramagnetic variety

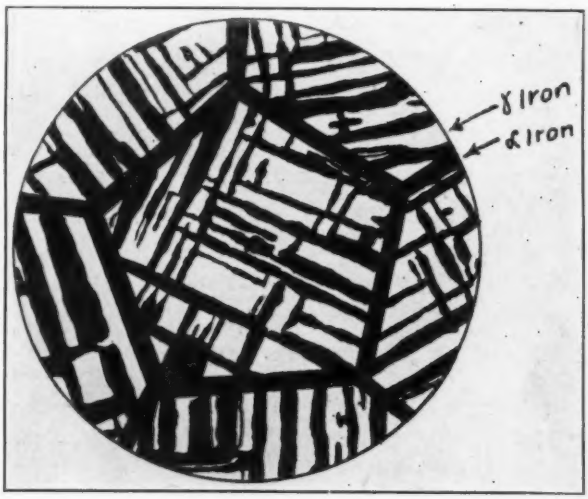

Fig. 2—Schematic Illustration of the Widmanstatten Type of Structure Assumed by Pure Iron in Passing from its Gamma to its Alpha Condition.

of body-centered iron should not be designated as beta iron, the name given to it by its discoverer, Floris Osmond. Whether ferromagnetic body-centered iron and paramagnetic body-centered iron are allotropically different depends, after all, on our definition of allotropy. We can readily define allotropy in such a way as to exclude the A_2 point, and this is done by most writers.

Let us now consider the mechanism of the gamma-alpha transformations in pure iron as it occurs at the A_3 point, when the iron passes from a face-centered to a body-centered cubic lattice (Fig. 1). Investigations carried on at Harvard University, chiefly by Dr. C. H. Chou have, I think, thrown a great deal of light upon this phenomenon. In a paper written in 1927[1], I expressed the belief that when a gamma grain transforms into one or more alpha grains the transformation does not occur abruptly through the entire mass of the gamma grain but gradually through the new alpha phase locating it-

[1]Albert Sauveur, "The Widmanstatten Structure of Iron Alloys," Bicentenary Number, *Proceedings*, American Philosophical Society, Vol. LXVI, 1927.

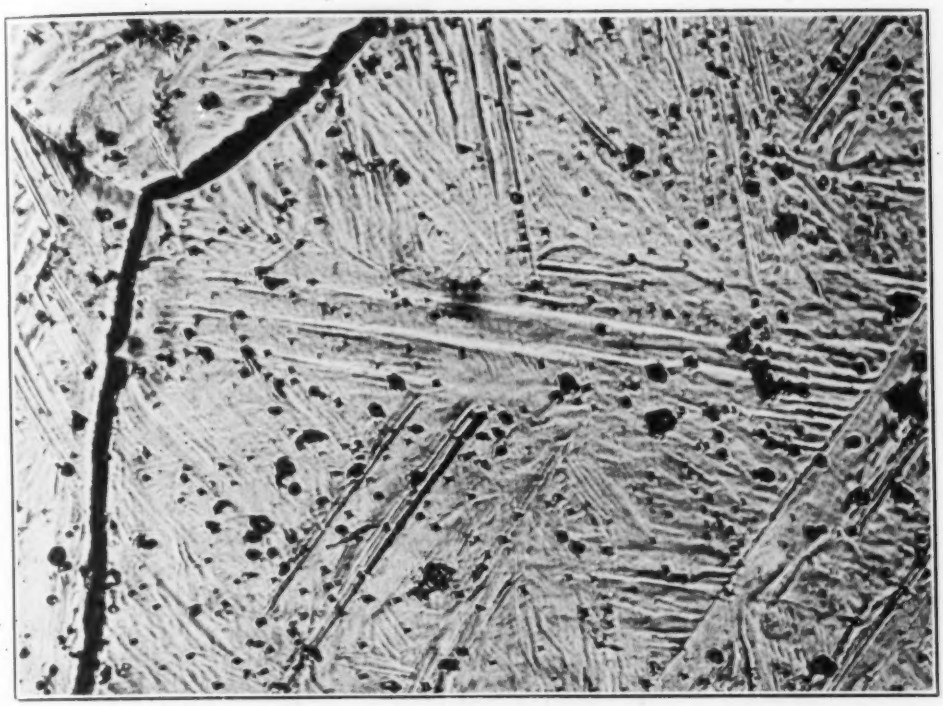

Fig. 3—Photomicrograph of Electrolytic Iron Heated in Vacuo at 1080 Degrees Cent. (1975 Degrees Fahr.) for 2 Hours and Quenched in Mercury.

self at the grain boundaries and along some of the crystallographic planes (probably the 1 1 1 planes) of the gamma grain, in such a way that if the gamma grain could be preserved incompletely transformed, a Widmanstatten type of structure would be revealed, as shown schematically in Fig. 2.

In order to ascertain, if possible, the correctness of this view, polished samples of electrolytic iron were placed in quartz tubes which were then evacuated, sealed and heated to 1000 degrees Cent. (1830 degrees Fahr.) for a period of two hours. The tubes were then plunged in ice cold mercury and quickly broken to permit the liquid to come into direct contact with the iron samples, causing their rapid cooling.[2] It was hoped by such means to produce and retain on the polished surfaces a pattern which would reveal the mechanism of the allotropic transformation. It will suffice to reproduce here (Fig. 3) one of the many photomicrographs taken to illustrate this point. It seems evident, as was anticipated, that the alpha phase

[2]Albert Sauveur and C. H. Chou, "The Gamma-Alpha Transformation in Pure Iron," American Institute of Mining and Metallurgical Engineers Technical Publication No. 169, Feb., 1929.

forms first along some of the crystallographic planes of the gamma iron and around the grain boundaries. Owing to the increased volume involved in the transformation, the alpha constituent stands in relief. The transformation then continues until the gamma iron is completely converted into alpha iron. The pattern indicative of the mechanism of the transformation, however, is retained impressed on the polished surface.

Having looked into the mechanism of the transformation of the solvent, let us now consider the mechanism of similar transformations as they occur in steels of various carbon contents. It will be in order to consider three types of steels, namely hypoeutectoid steel, eutectoid steel and hypereutectoid steel. It will also be in order to consider in each case the mechanism of transformation: (1) on slow cooling and (2) on rapid cooling—meaning by slow cooling a cooling so slow that the steel becomes pearlitic and by rapid cooling so rapid that the steel becomes martensitic.

Hypoeutectoid Steel Slowly Cooled. Let us consider hypoeutectoid steel containing 0.40 to 0.50 per cent of carbon cooling slowly through its thermal critical range. Above that range it consists, like all steels, of a solid solution of carbon in gamma iron (face-centered cubic). There is some dispute as to the exact position occupied by the carbon atoms in the solvent or by Fe_3C molecules, but it will not be necessary for our purpose to take part in it. The solid solution is called austenite. When this hypoeutectoid steel reaches its first critical point $Ar_{3\text{-}2}$, a new phase in the form of alpha iron (body-centered cubic) comes into existence at the boundaries of the austenitic grain and along some of the crystallographic planes, giving rise to a Widmanstatten type of structure (Fig. 4). This alpha iron at the instant of its formation may and probably does retain some carbon in solution, but on account of its slight solubility for carbon, the latter leaves it to concentrate in the surrounding austenite, which is capable of absorbing as much as 0.85 per cent carbon and which does finally reach that composition at the point Ar_1 690 degrees Cent. (1275 degrees Fahr.) when it transforms into pearlite. The alpha phase now practically free from carbon is called ferrite.

The transformation of austenite, now of eutectoid composition, into pearlite on slow cooling, results from the allotropic transformation of the solvent (gamma iron) into alpha iron which, being unable to retain carbon in solution, causes the separation or formation of the

Fig. 4—Widmanstatten Structure of Medium High Carbon Steel After Slow Cooling to Room Temperature. The White Needles are Ferrite (Alpha Iron) and the Dark Matrix, Pearlite. × 100. (C. H. Chou).

carbide Fe_3C (cementite). Ferrite and pearlite are thus the microscopical constituents of the Widmanstatten structure resulting from slow cooling of hypoeutectoid steel.

Hypoeutectoid Steel Rapidly Cooled. When the steel under consideration is cooled so rapidly that the decomposition of austenite is delayed until a temperature of some 300 degrees Cent. (570 degrees Fahr.) is reached, it remains incomplete and the steel is hardened. Its structure is then generally described as martensitic. The mechanism of the transformation, however, remains the same. The alpha phase separates along some of the crystallographic planes of austenite, forming the so-called martensitic needles and imparting to the metal the acicular structure characteristic of hardened steel. At the instant of its formation the alpha iron present in the needles probably retains carbon in solid solution. Alpha iron, however, having little solubility for carbon the latter tends to come out of solution. While the alpha needles are surrounded by austenite still able to absorb more carbon than it contains, time is now denied for the passage of carbon from the alpha needles into the gamma matrix, in which they are embedded, and this results in extremely fine particles of cementite being formed within and retained by the needles, converting them, in my opinion, into troostite. It is evident that the marten-

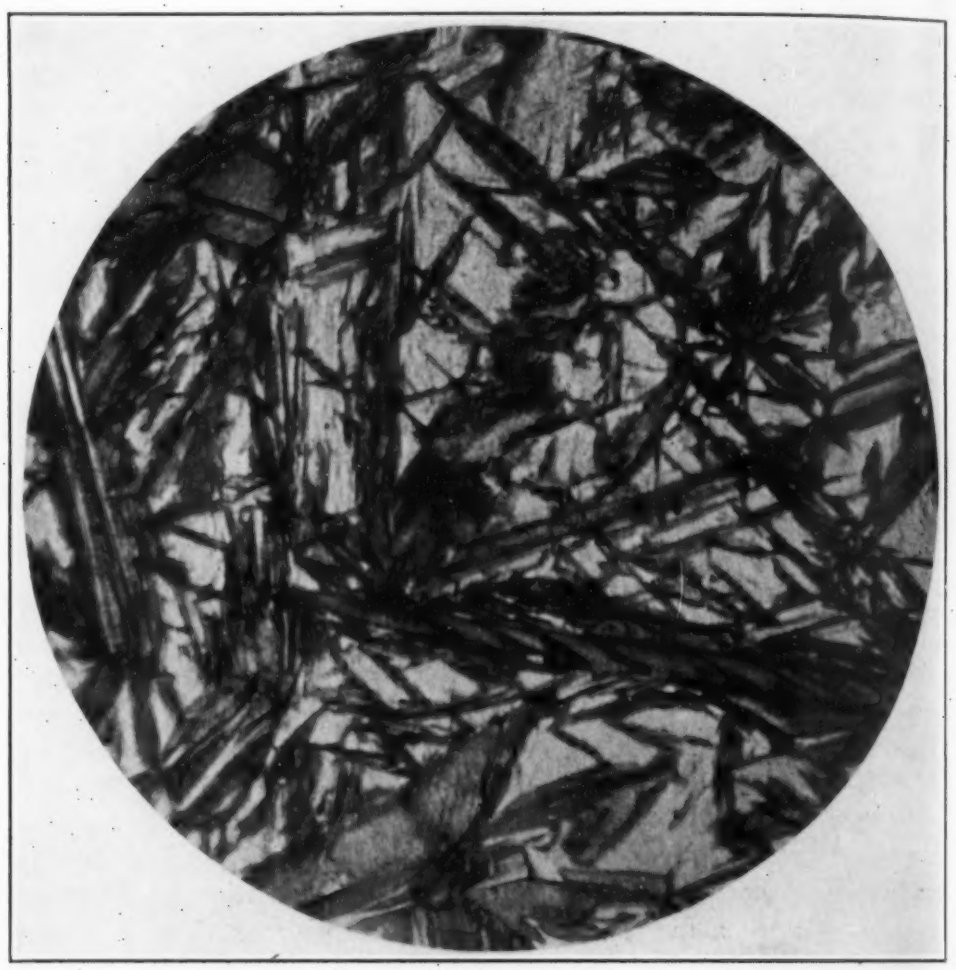

Fig. 5—Martensitic Structure. The Dark Needles are Troostite (?), the Light Matrix Undecomposed Austenite. × 2200. (Lucas).

sitic structure (Fig. 5) resulting from rapid cooling is similar to the Widmanstatten structure resulting from slow cooling, although the former is very much finer.

The constituents of the Widmanstatten structure resulting from slow cooling in hypoeutectoid steel are ferrite and pearlite, while the corresponding constituents of the fine Widmanstatten structure, generally known as a martensitic structure, resulting from rapid cooling are troostite and undecomposed austenite.

The similarity between the Widmanstatten type of structure and the martensitic type is well brought out by magnifying the latter to so high a degree that the constituents of both structures appear as having substantially the same dimensions. This has been done in the

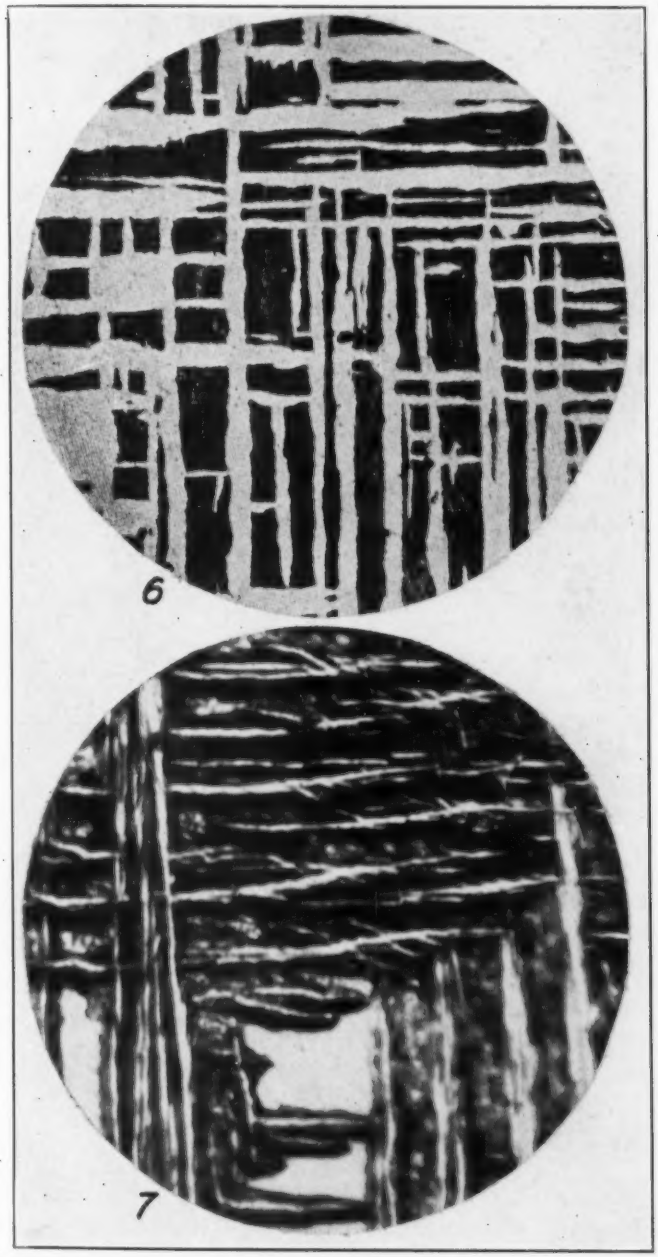

Fig. 6—Widmanstatten Structure in Steel Containing 0.55 Per Cent Carbon. × 30. (Belaiew).
Fig. 7—Martensitic Structure. × 3000. (Lucas).

photomicrographs of Figs. 6 to 9. Figs. 6 and 8 show the Widmanstatten structure of hypoeutectoid steel containing 0.55 per cent carbon under low magnification, while Figs. 7 and 9 represent the structure of hardened steel at high magnification.

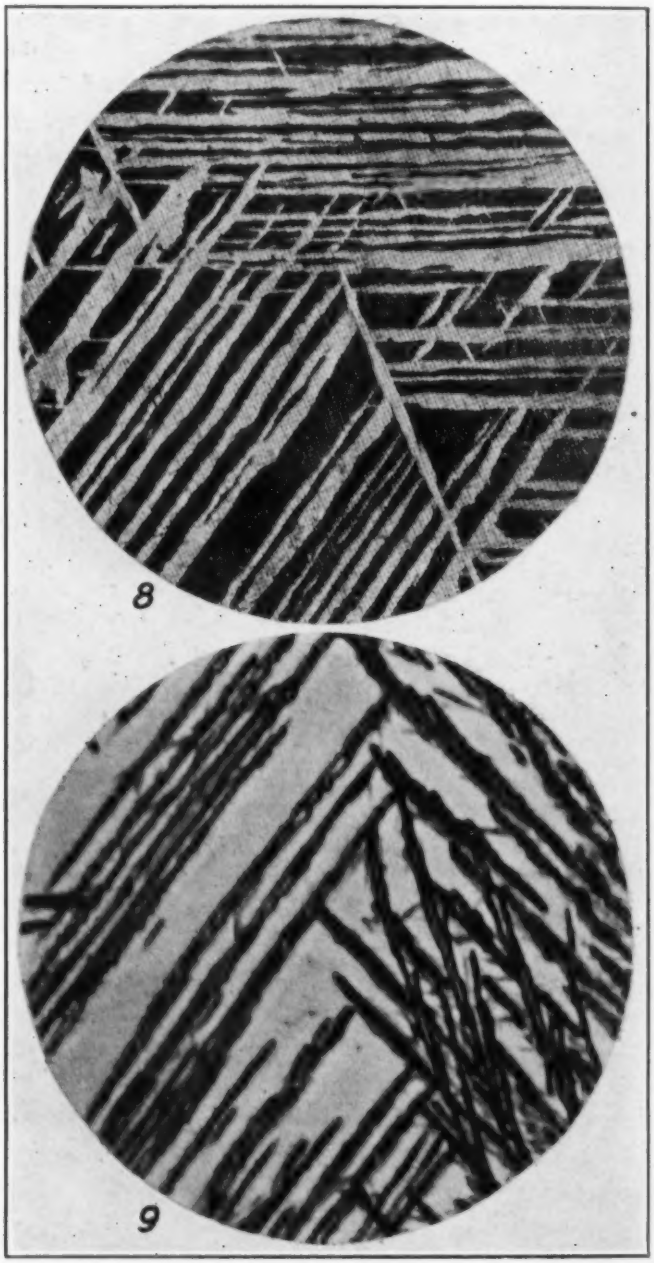

Fig. 8—Widmanstatten Structure in Steel Containing 0.55 Per Cent Carbon. × 30. (Belaiew).
Fig. 9—Martensitic Structure. × 3000. (Lucas).

Eutectoid Steel Slowly Cooled. When steel containing 0.85 per cent carbon is slowly cooled through its critical point $Ar_{3\text{-}2\text{-}1}$ at some 700 degrees Cent. (1290 degrees Fahr.), it is believed that the gamma-alpha transformation of the solvent starts at the boundaries and

Fig. 10—Widmanstatten Structure of Hypereutectoid Steel. The White Needles are Cementite, the Dark Matrix, Pearlite. × 100. (E. L. Reed).

along some of the crystallographic planes of the austenite grains. Were it possible to retain the steel but partially transformed into pearlite, a Widmanstatten structure would again be obtained with pearlite (the alpha phase) at the crystallographic planes embedded in undecomposed austenite. The carbon present in the alpha phase at the moment of its formation cannot migrate into the surrounding austenite, because the latter already contains all the carbon it can hold. The carbon in the alpha phase consequently separates into parallel plates of cementite giving rise to the formation of pearlite, made possible by slow cooling. This process continues until the whole of the steel has been converted into pearlite.

Eutectoid Steel Rapidly Cooled. When eutectoid steel is cooled sufficiently rapidly to be hardened, the austenite remains partially transformed but the mechanism of that transformation remains the same. The alpha phase forms at some of the crystallographic planes, giving rise to a martensitic structure. The alpha needles may retain some carbon in solution, but the greater part of it is probably present in the form of minute cementite particles. The steel is now an ag-

gregate of troostite needles (generally called martensite needles) embedded in undecomposed austenite.

Hypereutectoid Steel Slowly Cooled. When hypereutectoid steel cools slowly through and below its Ar_{cm} point, needles of cementite form along some of the crystallographic planes of austenite and around the boundaries, giving rise again to a Widmanstatten type of structure (Fig. 10). The mechanism of the formation of cementite from hypereutectoid austenite might be conceived as due to alpha iron forming in the usual manner but taking with it enough carbon (6.67 per cent) to be converted into the carbide Fe_3C. At the Ar_1 point the remaining austenite now of eutectoid composition is converted into pearlite, following the mechanism already described.

Hypereutectoid Steel Rapidly Cooled. Rapid cooling may prevent any separation of free cementite. When the delayed transformation takes place at some 300 degrees Cent. (570 degrees Fahr.) the hypereutectoid austenite is transformed into needles of troostite (called martensitic needles by most writers) embedded in undecomposed austenite. The troostite needles form along some of the crystallographic planes of the mother austenite, imparting to the hardened steel a structure resembling the Widmanstatten pattern although on a much finer scale.

Martensite versus Troostite. Although the structure of hardened steel has for many years, and still is by some, loosely described as consisting of martensite, it is evident that it is an aggregate made of two constituents, one in the form of "needles" embedded in another constituent constituting the matrix. There seems little doubt but that the matrix is undecomposed austenite. The needles are generally described as "martensitic" needles and are assumed by many to consist of a solid solution of carbon in alpha iron. While it is possible or even probable that some carbon may be retained in solid solution in the alpha iron of the needles, it seems also certain that they contain a large proportion of the carbon in the form of extremely finely divided particles of cementite. They are, therefore, of the nature of troostite and, in my opinion, should be called troostite. A martensitic structure would then refer to the fine aggregate of austenite and troostite, of which hardened steel is constituted. There seems to be some reluctance to consider the needles of hardened steel as consisting of troostite. Is not troostite, by definition, alpha iron containing finely divided particles of cementite?

It has been said that at the instant of their formation the needles

consist wholly of a solid solution of carbon in alpha iron and that the name of martensite might be preserved to designate this solution. It is agreed, however, that immediately martensite is formed cementite particles are precipitated, converting this solid solution into a fine aggregate answering to our definition of troostite, whether or not some carbon may be retained in solid solution. It has also been suggested that the name martensite might be retained for the needles in which cementite occurs in submicroscopic dimensions and troostite for the coarser aggregate in which they occur in microscopic dimensions. The distinction is too subtle to be acceptable. I prefer to retain the designation of martensite for the aggregate of troostite and austenite constituting hardened steels.

The Constituents of Iron-Carbon Alloys. Five solid phases are to be considered in alloys of iron and carbon, namely (1) gamma iron, (2) alpha or delta iron, (3) solid solution of carbon in gamma iron, (4) solid solution of carbon in alpha or delta iron, and (5) the carbide Fe_3C. Any alloy of iron and carbon, regardless of its composition, must of necessity be made up of one or of two phases when it is stable, according to the phase rule, or of three phases when it is unstable. Austenite is the solid solution of carbon in gamma iron (phase 3); Troostite is an aggregate of alpha iron (phase 2) or of alpha iron containing carbon in solution (phase 4) and of cementite (phase 5). Martensite is an aggregate of troostite and of austenite, hence of phases 3, 4 and 5, or of phases 2, 4 and 5, and hence unstable. Sorbite is an aggregate of alpha iron (phase 2) and of cementite (phase 5). It differs from troostite in greater coarseness of structure and in the probable absence of carbon retained in solution in alpha iron. Pearlite is an aggregate of alpha iron (phase 2) and of cementite (phase 5), in which both phases have crystallized in the form of distinct parallel plates. Ferrite is alpha iron (phase 2) and cementite the carbide Fe_3C (phase 5). Gamma iron (phase 1) cannot be retained at room temperature.

Hardness of Martensite

From the foregoing considerations the hardness of martensite containing a sufficient amount of carbon may be ascribed (1) to the fineness of its structure, (2) to the acicular character of its structure, (3) to the hardness of the troostitic needles which probably consist of a supersaturated solid solution of carbon in alpha iron and of numer-

ous finely divided particles of cementite, (4) to the hardness of the retained austenite which being supercooled is harder than normal, stable austenite and (5) to the existence of severe internal strains.

Tempering of Martensite

One of the constituents of martensite, the retained austenite being supercooled and the other, the troostitic needles, being supersaturated it may be anticipated that heating above room temperature will produce changes in both constituents. The supercooled austenite will tend to transform into troostite through gamma iron being converted into alpha iron. This may at first produce increased hardness because a supersaturated solid solution is harder than a supercooled solution. The troostitic needles will tend to expel more of the carbon which they may still contain in solution while the cementite particles will tend to agglomerate into masses of larger size, both of which phenomena will decrease the hardness.

Some Recent Views of the Hardening of Steel

Some explanations of the hardening of steel by quenching recently offered should be mentioned as they have been accepted by many. They may be described as (1) the grain-size theory, (2) the dispersion theory and (3) the distortion theory.

The Grain-size Theory. The grain-size theory first enunciated by Jeffries and Archer and called by them the "slip interference theory" ascribes the hardening of steel primarily to the presence of very small alpha grains, described by them as submicroscopic. While the diffuseness of the alpha lines of the X-ray diffraction patterns of quenched steel is claimed to be due to the existence of these small alpha grains, others believe that it is caused by the distortion of the space lattice resulting from the presence within it of carbon atoms or of very small particles of the carbide Fe_3C. According to the grain-size theory, were it possible to obtain alpha iron free from carbon in the form of submicroscopic grains it would be practically as hard as quenched high carbon steel. Jeffries and Archer also believe that the presence of minute particles of cementite increase the hardness of the martensite structure.

Dispersion Theory. The dispersion theory attributes the hardening of steel by rapid cooling to the precipitation of minute particles of cementite. Jeffries and Archer consider it to be a contributory

cause of the hardening of steel, others believe it to be the primary cause. This theory owes its existence to the behavior of duralumin on aging after quenching. It is well known that duralumin which is an alloy of aluminum containing 3 to 4 per cent of copper, forty-eight hours or more after it has been quenched in water from a temperature of some 500 degrees Cent. (930 degrees Fahr.) increases substantially in strength, hardness and ductility. This is believed to be due to the precipitation on aging of numerous small particles of the compound $CuAl_2$ which increases resistance to slip hence also strength and hardness. Why they should also increase the ductility has never been explained. Similar changes of properties have been observed in other nonferrous alloys on aging after quenching and attempts have been made to include within this theory the hardening of steel. It is further claimed that maximum hardness corresponds to a certain "critical" size of the separated particles. Any treatment such as the tempering of steel which causes larger particles to form results in decreased hardness.

Distortion Theory. Rosenhain and others ascribe the hardness of steel after quenching to a distortion of the space lattices of the alpha crystals resulting from the presence within them of carbon atoms or of precipitated iron carbide in a very fine state of division, distorted space lattices imparting to the crystals greater resistance to slip, hence greater hardness. The diffuseness of the alpha lines of the diffraction patterns of hardened steel is offered as an evidence of distorted space lattices. It has already been noted that some believe this diffuseness to be due to the extremely small size of the alpha grains. On tempering hardened steel the alpha lines become more distinct, this being due according to the distortion theory to the removal of lattice distortions, and according to the grain-size theory to the formation of larger alpha grains.

In our zeal to discover fundamental laws which explain as many apparently related phenomena as possible, there is danger of covering too great a territory. While the dispersion or the distortion theory may afford a satisfactory explanation of the hardening of some nonferrous alloys after quenching, followed by aging, it is well to remember that the behavior of steel is in many important respects radically different from the behavior of these nonferrous alloys. The degree of hardness imparted to steel by rapid cooling is of a degree of magnitude altogether unrelated to the increased hardness observed in these nonferrous alloys. Steel acquires its hardness in the

quenching bath and not on subsequent aging or at least not to any appreciable degree. The attempt to reconcile the behavior of steel with that of some nonferrous alloys by claiming that steel ages in the quenching bath fails to carry conviction. If the hardening of steel is due to the dispersion of minute particles of Fe_3C, the phenomenon must take place almost instantaneously in the quenching bath and not leisurely at room temperature as is the case with the precipitation of $CuAl_2$ particles in duralumin. The hardening of steel depends upon the allotropic transformation of the solvent passing in part at least from the condition of gamma iron to that of alpha iron. If the solvent did not undergo an allotropic transformation, steel would not harden. Nothing of a similar nature is observed in nonferrous alloys, the solvent, aluminum in the case of duralumin failing to undergo any allotropic transformation.

The ductility of medium high or high carbon steel is sharply decreased by hardening, while in marked contrast, the ductility of duralumin is increased by quenching followed by aging. It is not easily understood why the dispersion of a hard constituent would in the case of steel increase the hardness and decrease enormously its ductility, while in duralumin and some other alloys it increases both the hardness and ductility.

The Following Questions are Still Pertinent

(1) What direct evidence have we of the existence of submicroscopic grains of alpha iron in hardened steel?

(2) Granting their existence what direct evidence have we that they explain satisfactorily the hardening of steel by quenching?

(3) What direct evidence have we of the presence in hardened steel of numerous cementite particles of submicroscopic size?

(4) Granting their existence, what direct evidence have we that they explain satisfactorily the hardening of steel by quenching?

(5) What direct evidence have we that maximum hardness corresponds to a certain "critical" size of the separated cementite particles?

(6) What causes the diffuseness of the alpha lines in the diffraction patterns of hardened steel: (a) smallness of grains, (b) carbon atoms within the space lattices or (c) minute carbide particles within the lattices?

In spite of many efforts to gather steel within the fold it still

remains very much in a class by itself and it may be the part of wisdom to wait a while longer before joining the followers of the Metallurgical Pied Pipers.

Appendix

R. H. Aborn

Although comparatively few quantitative investigations of the changes in lattice dimensions of iron at elevated temperatures have been made, it has seemed feasible to correlate existing data with a twofold aim:—

(1) To show the volume (or density)—temperature relationships at all temperatures below the melting point.

(2) To compare the different modifications of iron with respect to these properties.

The only pertinent data published are those of Westgren and Wever. In addition, Hull's data on alpha iron at room temperature are employed.

Table I presents this data in condensed form, together with calculations as follows:—

Table I

Phase	Temp. °C	Lattice Type B. C. C. = Body centered cubic F. C. C. = Face centered cubic	Number of atoms per unit cube	Length of cube edge Å (1Å. = 10^{-8}cm.)	Distance of closest approach of atoms Å	Relative volume per Fe atom (cubic Å)	Calculated Density
αFe	20°	B. C. C.	2	2.86 (Hull)	2.48	11.70	7.85
$\alpha(\beta)$Fe	800°	B. C. C.	2	2.90 (Westgren)	2.51	12.19	7.53
$\alpha(\beta)$Fe	900°—	B. C. C.	2	2.905 (from graph)	2.515	12.26	7.49
γFe	900°+	F. C. C.	4	3.60 (Wever)	2.55	11.66	7.86
γFe	1100°	F. C. C.	4	3.63 (Westgren)	2.57	11.96	7.67
γFe	1400°—	F. C. C.	4	3.68 (Westgren)	2.60	12.46	7.37
δFe	1400°+	B. C. C.	2	2.93 (Westgren)	2.54	12.58	7.30

Closest (Mutual) Approach of Atoms

A. In the body-centered cubic lattice this occurs along a cube

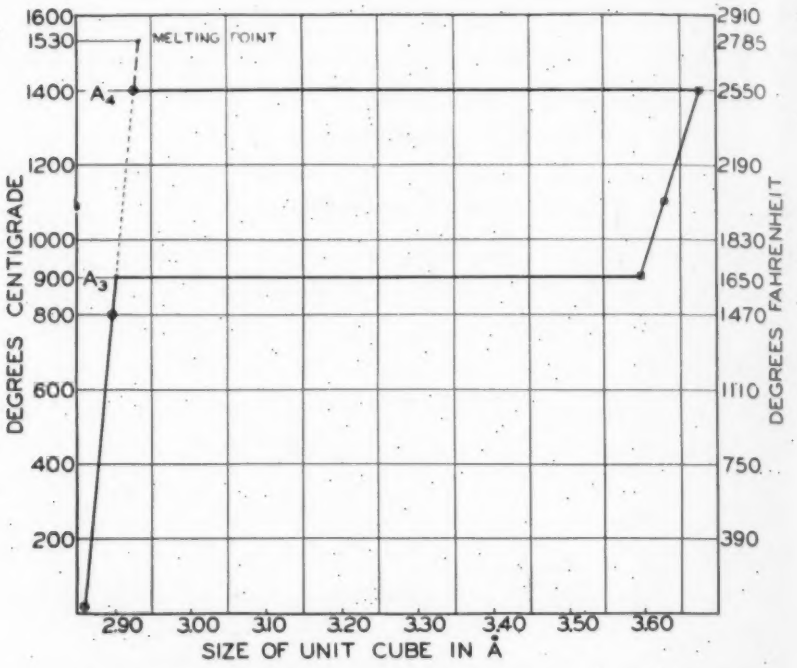

Fig. 11—Relationship Between Cube Edge Length and Temperature from Which was Obtained the Value for Alpha Iron at 900 Degrees Cent. (1650 Degrees Fahr.)

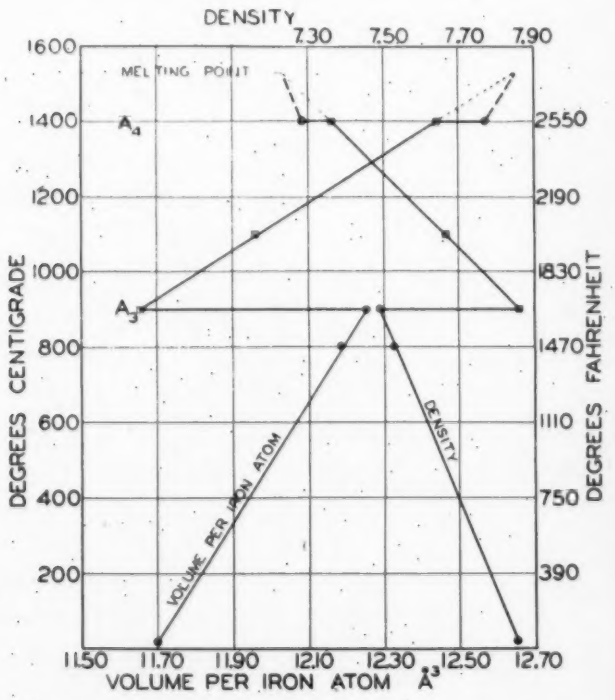

Fig. 12—Density, Volume and Temperature Relationships in Iron.

diagonal, where the center atom is equally near eight other atoms (the corner atoms). If a = cube edge length,

$$\text{Distance of closest approach} = \frac{a\sqrt{3}}{2}$$

B. In the face-centered cubic lattice this occurs along a face diagonal where the face-centered atom is equally near twelve other atoms (four at the corners of the face, three at the other face centers, and three at corresponding face centers of the cube adjacent). If a = the cube edge length,

$$\text{Distance of closest approach} = \frac{a\sqrt{2}}{2}$$

Relative Volume per Iron Atom

(Note: This is not to be construed as the volume occupied by the atom, but simply as a relative measure of the available space per atom, computed as follows):

A. In the body-centered cubic lattice,

$$\text{Relative volume per atom} = \frac{\text{volume of unit cube}}{2}$$

since there are two atoms associated with each unit cube, (one at the center plus eight at the corners (each of which is shared by eight cubes); thus $1 + 8(\frac{1}{8}) = 2$.

B. In the face-centered cubic lattice,

$$\text{Relative volume per atom} = \frac{\text{volume of unit cube}}{4}$$

since there are four atoms associated with each unit cube [one at the center of each of the six faces (therefore shared by two cubes) and one at each of the eight corners (therefore shared by eight cubes)]; thus

$$\frac{1 \times 6}{2} + \frac{1 \times 8}{8} = 4$$

Density

$$\text{Density} = \frac{\text{mass}}{\text{volume}} = \frac{(n)\ (A)\ (m')}{\text{volume of unit cube}}$$

where n = number of atoms associated with each unit cube,
A = atomic weight of iron,
m′ = absolute mass of an atom of unit atomic weight
= 1.643 x 10^{24} gm.

In Fig. 11 is plotted the relationship between cube edge length and temperature, from which was obtained the value for α (β) iron at 900 degrees Cent. (1650 degrees Fahr.). It shows clearly that the same approximately linear function relates thermal expansion and temperature for all the body-centered cubic phases, while a different approximately linear function applies to the face-centered cubic phase.

Fig. 12 represents the density (and volume)—temperature relationships. The identity of the three body-centered phases is again apparent. It is interesting to note that the face-centered cubic lines when extrapolated beyond the A_4 point (as dotted lines) intersect the body-centered cubic lines at the melting point of iron.*

DISCUSSION

Written Discussion: By Dr. Kotaro Honda, Tohoku Imperial University, Sendai, Japan.

Dr. Sauveur's paper contains many points of interest, but his main argument seems to lie in his new view that martensite obtained by quenching carbon steels consists mainly of troostite, putting for a while retained austenite[1] out of consideration. In other words Dr. Sauveur assumes martensite to be a structure in which fine cementite particles are dispersed in ferrite ground, that is, to be troostite in the real sense; while martensite is usually assumed to be a supersaturated solid solution of carbon in alpha iron. He makes, however, some allowance to his view in making statements, such as "martensite is a solid solution containing also finely divided cementite particles," (page 199), or "troostite, whether or not some carbon may be retained in solid solution" (page 211), etc.

In order to avoid ambiguity in this argument, let us adopt the following definitions:—

(a) Austenite is a structure in which carbon is contained in gamma iron as a solid solution.

(b) Martensite is a structure in which carbon is contained in alpha iron as a solid solution.

(c) Troostite is a structure in which fine particles of cementite is dispersed in ferrite ground.

Then according to Dr. Sauveur, quenched steel consists of troostite

*Dr. H. H. Lester, research physicist, Watertown Arsenal, first suggested this as an approximate check on the validity of the data.

[1]By regulating the velocity of rapid cooling, the amount of retained austenite can be made very small or zero.

and retained austenite, with or without a small quantity of martensite, while according to the prevailing view, it consists mainly of martensite, with or without some retained austenite. Again, after Dr. Sauveur, a structure which is obtained by tempering a quenched steel at 300 degrees Cent. (570 degrees Fahr.), consists of a troostite of a larger grain size or a mixture of martensite and cementite (page 211), whereas in the usual view is consists solely of troostite. The essential point of difference of Dr. Sauveur's view is therefore the annihilation of martensite in the sense above defined, or at least gives this structure a small importance; because after him, in quenched steel martensite is, in its greater part, decomposed into troostite as soon as it is formed.

The argument by which he was led to this view, is mainly based on microscopic investigation, but it will be noted that to understand the true nature of quenched steel, it is very important to study its nature from different viewpoints of investigation, that is, from the viewpoints of thermal change, of thermal dilatation, of electric resistance, of magnetic change, of X-ray analysis, etc. Each of these methods of investigation has its own merit and demerit, and gives a new information regarding its nature. Thus during tempering a quenched steel[2], electric resistance decreases in two steps, that is, in the vicinity of 150 and 270 degrees Cent. (300 and 520 degrees Fahr.) this decrease amounting to 100 per cent in the case of eutectoid steel; as the electric resistance of a solid solution is generally greater than that of the mixture, the presence of this latter step[3] shows that the solid solution of martensite rapidly decomposes in the range of 270 degrees Cent. (520 degrees Fahr.) into iron and cementite. On the other hand, if troostite be the structure of the quenched steel, this large decrease of electric resistance must be explained by the coagulation of cementite particles, but as the effect of grain size on the electric resistance is very small, the above decrease of resistance cannot be explained by the coagulation of cementite. The magnetization of a quenched steel increases by tempering also in two steps, this increase amounting to 30 per cent in the case of eutectoid steel. If troostite be the principal constituent of the quenched steel, this increase of magnetization by tempering must be attributed to the coagulation of cementite particles, but as the greater part of magnetization owes its origin to ferrite and only a small part to cementite, it is very difficult to explain the large increase of magnetization by the coagulation of cementite particles. These facts confirm that quenched steel consists mainly in a solid solution of carbon in iron and rapidly decomposes in the vicinity of 270 degrees Cent. (520 degrees Fahr.). X-ray spectral lines of martensite are much diffused as compared to those of annealed steel and become sharp on tempering at 300 degrees Cent. (570 degrees Fahr.); this also indicates that the space lattice of martensite is distorted by the presence of carbon atoms in the lattice, and that by the tempering carbon atoms are set free and consequently the distortion of the lattice vanishes.

Dr. Sauveur states that during quenching the first decomposition

[2, 3] Kotaro Honda, "On the Nature of Martensite Crystals," TRANSACTIONS, American Society for Steel Treating, Vol. 16, 1929, p. 97.

product of austenite is a solid solution of carbon in iron, but as soon as it is formed, the dissolved carbon is precipitated as cementite; but it must be noticed that by a rapid cooling, the decomposition of austenite takes place near the room temperature and hence the precipitation of carbon can hardly take place; because this process involves the movement or diffusion of carbon atoms through iron lattice; thus carbon atoms will remain very probably within the lattice as a supersaturated solid solution.

From the facts so far enumerated, it is to be concluded that contrary to Dr. Sauveur's view, a quenched steel consists of a solid solution of carbon in alpha iron, that is, of martensite, but not a mixture of ferrite and dispersed cementite particles, that is troostite, putting some retained austenite out of consideration.

The following points in his paper, which are not so important, may deserve a notice:—

(1) Volume change of iron at A_3 and A_4 points R. H. Aborn's calculation for the volume change at these two transformation points gives too large a value as compared with those actually observed by S. Sato[4] in this institute. The lattice constants used by Aborn are probably not quite correct; his results of calculation are as follows:—

$$\frac{\delta v}{v} \text{ at } 900^\circ \text{ C in the } \alpha \rightarrow \gamma \text{ change} = 5.0 \text{ per cent,}$$

$$\frac{\delta v}{v} \text{ at } 1400^\circ \text{ C in the } \gamma \rightarrow \alpha \text{ change} = 0.96 \text{ per cent.}$$

S. Sato also calculated the same change and obtained

$$\frac{\delta v}{v} \text{ in the } \alpha \rightarrow \gamma \text{ change} = 0.0082 \text{ per cent,}$$

$$\frac{\delta v}{v} \text{ in the } \gamma \rightarrow \alpha \text{ change} = 0.0023 \text{ per cent;}$$

while his directly observed values are respectively 0.0084 and 0.0025 per cent; the agreement between these two is satisfactory.

(2) Change of magnetization by heating or cooling and the corresponding heat evolution or absorption.

Dr. Sauveur believes that during heating and cooling the magnetic change occurs suddenly or abruptly; but this is not actually the case. In a strong field, the magnetization during heating and cooling is quite reversible, and as the temperature is raised, it begins to diminish from room temperature, its rate of decrease becoming always greater, till the magnetization vanishes at 790 degrees Cent. (1455 degrees Fahr.). This change of magnetization is intimately connected with the thermal change observed during heating and cooling. If the field is weak, the change of magnetization is much complicated by the effect of thermal agitation on magnetization.

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[4]S. Sato, "Dilatometric Investigation of the A_3 and A_4 Transformations in Pure Iron," Science Reports, Tohoku Imperial University, Sendai Japan, Vol. 14, No. 5, 1925, p. 513.

The explanation of these changes of magnetization and of the true nature of the magnetic or A_2 change required too much space to be given here and hence are merely referred to my book "Magnetic Properties of Matter" (see § 28, 31, 32, 57).

Written Discussion: By A. Portevin and P. Chevenard, Paris, France.

In the course of our investigations on the tempering of hyperquenched special steels, we observed three types of decomposition of carburized austenites[5]:

1. During heating, precipitation of the carbon in the form of complex carbide, and at the same time a return of the iron to the alpha state; the "ferrite-carbide" aggregate, stable in the cold, being restored, no transformation takes place on cooling.

2. Precipitation of a part of the carbon in the form of carbide during the heating; on cooling, the austenite, deprived of a part of the stabilizing elements, undergoes the quenching transformation that gives rise to martensite (secondary quenching).

3. Precipitation of part of the carbon in the form of carbide during the heating, without decomposition of the austenite on cooling; this is the case with certain ferronickels to which carbon has been added.

These phenomena are shown clearly by the differential dilatometric method, which is particularly advantageous for the study of very slow reactions and isothermal transformations.

As in every destruction of a false equilibrium the rate of precipitation of carbide increases rapidly with the temperature, the intensity of the phenomenon observed during the course of tempering depends, therefore, on the rate of heating and the temperature that is reached. At ordinary temperature the reaction velocity, although very low, is appreciable by the dilatometric method, as is proven by the experiences of Mr. Guillaume on the instability of carburized ferronickel[6].

The three types of reactions are shown in the two accompanying diagrams, which relate to superhardened steels with small additions or special elements. The tempering cycles have been drawn while recording the expansion by the differential dilatometer: scale for pure iron; rate of heating equal to 250 degrees per hour.

Fig. 1. Cycles 1, 2 and 3: Weak precipitation of carbide, indicated by a slight contraction. Cycle 4: Explosive decomposition of the austenite with the liberation of a large amount of heat (bend A). The return of the iron to the alpha state is shown by the quasi-horizontal direction of the cooling curve, and the precipitation of complex cementite by the abnormality C. The Brinell hardness increases from 200 to 380 kilograms per square millimeter. Cycles 5, 6 and 7: No transformation other than a progressive lowering of the temperature of the abnormality C; the carbide

[5] P. Chevenard, "Dilatometric Analysis of Alloys, with Special Reference to Cast Iron," *Foundry Trade Journal*, July 1924.
P. Chevenard, "Experimental Researches on the Iron Alloys of Nickel and Chromium," Works and Memoirs of the International Bureau of Weights and Measures, Vol. 17, 1927, p. 88.

[6] Guillaume, "Metrological Researches on Nickel Steels," Works and Memoirs of the International Bureau of Weights and Measures, Vol. 17, 1927, p. 161.

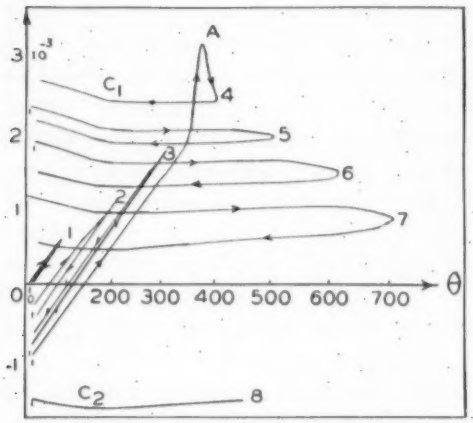

Fig. 1—Manganese Steel with a Carbon Content of 1.50 Per Cent and a Manganese Content of 2.05 Per Cent.

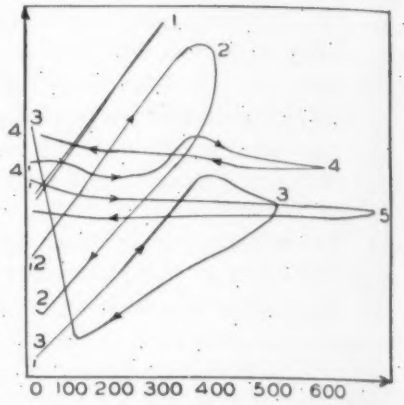

Fig. 2—Chromium Steel with a Carbon Content of 1.50 Per Cent and a Chromium Content of 2.25 Per Cent.

becomes richer in manganese at the expense of the ferrite. The hardness decreases because of the coalescence.

Fig. 2. Cycles 1 and 2: Precipitation of carbide, shown by a contraction. Cycle 3: Precipitation of carbide during the heating; on cooling, at 100 degrees Cent., "austenite to martensite" transformation, shown by an abrupt rise in the curve. The hardness increases from 200 to 400 kilograms per square millimeter.

Cycle 4: Decomposition of the martensite during heating indicated by a peculiarity of tempering.

Cycle 5: No transformation.

The reactions of the first two types are therefore quite different, although the increases in hardness are of the same order of magnitude.

In the first case, there is a return to the physical-chemical state of equilibrium, and the hardening is due to the formation of a very fine aggregate: structural quenching. In the other, there is the formation of a very hard constituent, martensite, outside the equilibrium physical-chemical quenching.

These two methods of tempering are not only characteristics of certain types of steels; they may occur in the same steel, depending on the thermal conditions under which the tempering is done, and they also superpose themselves in the industrial treatment of certain special steels, such as the high speed tool steels.

Excerpt from a Communication: By Dr. Harry Brearley, Torquay, England.

I admire, as always, the clearly reasoned statements Dr. Sauveur makes and the modest conclusions he draws from them. My own tentative views are in line with those expressed by Dr. Sauveur. As to an "explanation" of the hardening of steel I have nothing to say except that, as an assumption, I find the distortion theory the most useful aid to thought; this preference may, however, be entirely due to the fact that

I have been obliged for many years to think of steel in terms of its mechanical properties.

I notice, on page 213, that Dr. Sauveur speaks of "aging" as though it were confined to nonferrous alloys. Unfortunately something similar occurs in steel, and is occasionally of industrial importance and may lead to serious mishap. The ductility of steels becomes less on aging, the tensile strength is increased and the toughness (Izod value) is enormously lessened.

Excerpt from a Communication: By F. C. Thompson, the Victoria University of Manchester, England.

I am obliged to the American Society for Steel Treating for the opportunity of discussing Professor Sauveur's paper which I consider to be of real merit. With many of his conclusions I find myself in wholehearted agreement and he has, in my opinion, presented his case in a very fair manner. The points on which I am unable to agree, or on which fuller information is desirable, are the following:

(1) The proof that the Widmanstatten structure can be produced in pure iron is extremely interesting and is in line with the analogous structure in zinc and tin. It is mentioned, however, on page 204 that the alpha phase forms first around the grain boundaries. The evidence for this conclusion is not given unless it be the dark boundary shown in Fig. 3. This, however, would appear to be merely the "trench," due to volatilization, which is normal in samples which have been vacuum annealed.

(2) Professor Sauveur's view that the needles of "martensite" in a quenched steel are in reality troostite is one to which I find it difficult to subscribe. He is right, however, I believe in his view (page 212) that supercooled austenite on tempering is transformed directly to troostite without an intermediate martensitic stage. The statement, however, on page 210 "it seems also certain that they (i. e., the needles) contain a large proportion of carbon in the form of finely divided particles of cementite" is, so far as I am aware of the evidence, a rather sweeping one. It would be a valuable addition to his paper if Professor Sauveur would supply the reasons for his opinion in this respect. His new conception of martensite as an aggregate of troostite and austenite must stand or fall on this point.

(3) I am very glad too that Professor Sauveur has emphasized the point, which I believe to be fundamentally sound, that some austenite is retained in a normal quenched steel. Further I cannot but believe that he has been right in disregarding completely the view put forward in some circles that the change from austenite to pearlite is, in both directions, completed through the intermediate martensitic stage. I also am completely in accord with the author's opinion on page 211, that the view that a constituent containing cementite in an ultramicroscopic form should be termed martensite and one with microscopically resolvable cementite, troostite is, if only from the historical point of view, quite unacceptable.

(4) I am rather surprised that Professor Sauveur nowhere refers to the possibility of twinning playing a part in the formation and properties

of martensite. It does in the end, I admit, boil down nearly to his explanation (5)—page 212—of "severe internal strain," but it may be possible on a twinning hypothesis to explain in detail much which is otherwise rather obscure. Edwards and Carpenter first pointed out this possibility and the writer and W. E. W. Millington later were driven to give it further consideration *(Journal,* Iron and Steel Institute, 1924, No. 1). Our view was in essence that under the influence of the quenching stresses, austenite tended to twin, but that, since that phase is unstable below Ar_1, once the atomic movement necessary for this to take place was initiated the atoms would tend to reorientate themselves entirely in a stable form and thus become alpha iron. In the "twinned" bands so produced the carbon must of necessity, we think, still remain in highly supersaturated solution. It might also be worth pointing out that the very beautiful photomicrographs of Lucas—lately confirmed by Whitely—show very definite markings and "back-bone" in the martensite needles. Now, although this has, so far as I know, never been adequately explained, such "back-bones" are far more readily understood in a twinned constituent than in an almost isotropic phase such as troostite is generally believed to be. In my view, therefore, it is as yet too early to exclude the possibility of twin phenomena playing a part entirely from consideration.

(5) There is one further point in the paper proper to which I should like to draw attention, this time a question of fact. On page 209 it is stated that "were it possible to retain the steel but partially transformed into pearlite, a Widmanstatten structure would again be obtained with pearlite (the alpha phase) at the crystallographic planes embedded in undecomposed austenite." I have quenched 0.2 and 0.9 per cent carbon steels on many occasions during the course of the Ar_1 change. In every instance the pearlite is found in rounded patches which are free from austenite (transformed of course into martensite) and which show no indication of the crystallographic structure of the pre-existing phase.

(6) There is one further point connected with the appendix by Dr. Aborn which deserves reference. It is shown in Fig. 12 that extrapolation of the density—temperature curves for the face-and-body-centered lattices from the A_4 point leads to their meeting at the meeting point. On page 218 the footnote apparently suggests that this should be expected since the suggestion was made that such an agreement would supply an approximate check on the data. It would be of great theoretical interest to know why this should have been expected and why, if it be really true, it should not be regarded merely as a remarkable coincidence. So far as can be gathered from Fig. 12, the density of iron at the melting point is about 7.24. Now Berlin (Carnegie Scholarship Memoirs *Journal*, Iron and Steel Institute, 1926, XV, p. 1), whose work seems to me to be the most reliable, gives a value of the density of molten iron at that temperature of 7.40 or just under. The solid phase at the same temperature should be more, rather than less, dense and the discrepancy between this figure and that deduced from Fig. 12 needs explaining.

This discussion has greatly erred on the side of long windedness but

the importance of Professor Sauveur's paper seemed to deserve the most serious, even if rather extended, consideration.

Written Discussion: By O. W. Ellis, Director of Metallurgical Research, Ontario Research Foundation, Toronto, Canada.

There can be no question that the attempt here made by Dr. Sauveur to show that the mechanism of the decomposition of austenite is substantially the same, whether the resulting structure is pearlitic or martensitic, merits the attention of all interested in the study of the structure of iron and steel. However, though many may agree with him in regard to the manner in which the alpha phase tends to locate itself at the grain boundaries and along the crystallographic planes of the mother austenite under certain conditions of cooling, as many will find themselves in disagreement with him in regard to other of the points raised by him in his paper, which quite clearly has been prepared with the view of stimulating discussion of this most fascinating subject.

On reading this paper for the first time the opening paragraph created no special disturbance in the writer's mind. So long as we remain uncertain as to the nature of hardened steel, so long are we justified in experimenting and theorizing, and vice versa. On the preliminary reading of this paper the first sentences of the second paragraph, however, brought the writer up with a jerk. What if austenite did fail to transform in cooling through its thermal critical range? Would it have been so vitally important to civilization? Would the A.S.S.T. never have come into existence? Would all grades of iron and steel have remained nonmagnetic?

Somehow one feels that the author has jumped a little too far in his conclusions here. To take only the first of these vital sentences—"if austenite failed to transform in cooling through its thermal critical range." Let us suppose that it did fail so to do. How would this be likely to affect the equilibrium of the iron-carbon system?

As far as one can see, all that one need do to obtain an answer to this question is to assume the nonexistence of the A_3 point. Under these conditions, then, the iron-cementite diagram would take the form shown in Fig. 1 of this discussion. Those portions of the diagram which represent the equilibrium of these alloys at temperatures in excess of the A_3 point of pure iron would in all probability be unaltered. The A_3 line, however, would disappear, the eutectoid transformation would be eliminated and pearlite would therefore be unknown. But, so far as one can judge, the solubility of carbon in iron would remain unchanged at temperatures in excess of about 700 degrees Cent. (1200 degrees Fahr.) and could, at temperatures below 700 degrees Cent., be represented by such a solubility line as that denoted by the letters BC in Fig. 1.

As to the magnetic change point, there seems no valid reason why it should disappear, if the view is accepted that the change in internal energy, which makes itself evident by a marked evolution and absorption of heat, is due to "some change in the structure of the atoms themselves" (line 5, page 202, of Dr. Sauveur's paper).

So then, "if austenite failed to transform in cooling through its

thermal critical range," one would obtain, in place of the steels with which we are now so familiar, a series of alloys capable of having a secondary hardness developed within them by aging after quenching from temperatures in excess of those corresponding to the solubility line ABC of Fig. 1. Who is there who would dare to deny the possibility of the discovery having been made in early times that these common alloys of iron and carbon could be hardened by quenching and aging, had the equilibrium diagram for this system been as shown in Fig. 1?

Speculation as to the degree of hardness that could be developed in these alloys by the precipitation of cementite in sub-microscopic form I

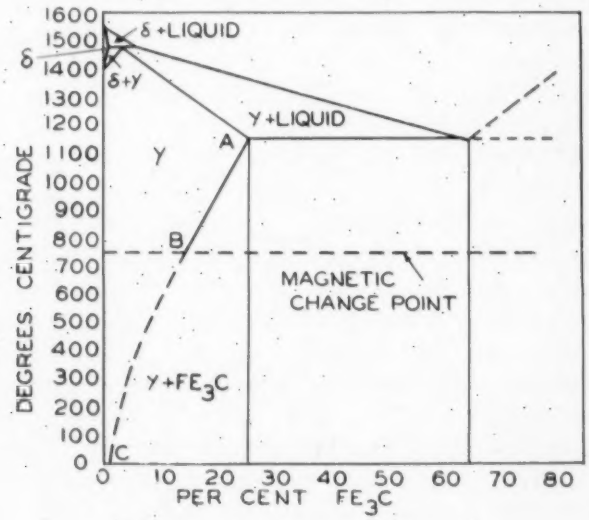

Fig. 1—Iron-Cementite Diagram.

leave to those who care to follow up this line of thought. It may be well to note, however, the similarity between this diagram (Fig. 1) and those of the iron-molybdenum and iron-tungsten systems and, further, to observe that the solubility of molybdenum in iron (24 per cent) in terms of the compound Fe_3Mo_2 is 45 per cent, that of tungsten in iron (33 per cent) in terms of the compound Fe_3W_2 is 48.1 per cent, while that of carbon in iron (1.7 per cent) in terms of the compound Fe_3C is 25.5 per cent—it is in these terms, among others, of course, that one has to think of the effect of these compounds on the hardness of iron.

Turning now to the suggestions made by Dr. Sauveur that the needle-shaped constituent seen in what has been so far universally referred to by the well-versed metallographist as a martensitic structure (not martensite) should be referred to as troostite, might it not be well if, before listing to his pipe (though the tunes he plays are well worth the hearing), one referred to the glossary of terms agreed upon by the Committee on the Nomenclature of Metallography, appointed by the Iron and Steel Institute in 1901, whose report, published in 1902 (*Journal*, Iron and Steel Institute, 1902, Vol. 61, p. 90), is still well worthy of note,

seeing that we have advanced but little in the study of the microstructure (per se) of steel since the time this glossary was first presented to the world.

Consider first the definition of austenite given in this glossary:

"AUSTENITE.—(Ger. Austenit, Fr. Austenite—Osmond.) (See Osmond's 'General Method for the Micrographical Analysis of Carbon Steels,' Photos. Nos. 10 and 11.)

This is described by Osmond as a constituent of steel softer and less magnetic than martensite, with which it is often associated. It is produced by quenching small sections of steel containing more than 1.5 per cent carbon in ice-cold water from a temperature of 1100 degrees Cent.

The structure of such quenched steel is developed by polishing on parchment moistened with liquorice solution or nitrate of ammonia after reheating to 200 degrees Cent., or by etching with hydrochloric acid in an electric current. The austenite remains white, the martensite becomes brown, and appears in zig-zag form. A sewing-needle drawn across the section scratches the austenite more deeply than the martensite."

Two constituents are here referred to—one, austenite, which remains white under certain conditions of polishing and etching and is softer and less magnetic than the other; the other, martensite, which becomes brown on etching and appears in a zig-zag form. The outstanding features of the latter constituent then are—

(a) its hardness,

(b) its high magnetic permeability,

(c) its zig-zag structure.

Consider next the definition of martensite given in this glossary:

"MARTENSITE.—(Ger. Martensit, Fr. Martensite—Osmond.) See Osmond's 'General Method for the Micrographic Analysis of Carbon Steels' *(L'Etude des Alliages)*, Photo. No. 8.

'A good example is afforded by a steel with 0.45 per cent carbon heated to 825 degrees Cent. and quenched at 720 degrees Cent., after slow cooling, in a refrigerating mixture at —20 degrees Cent. The structure is only shown vaguely by polishing in bas-relief, from which it is shown that the hardness of the mass is nearly uniform; but it is developed better by polish attack. (Photo. No. 8, 1000 diam.) We then see groups of needles or rather of rectilinear fibres oriented in parallel directions, and separated or not by granular or vermicular matter. Three systems of fibres respectively parallel to three sides of a triangle, cutting frequently in the same region and characterizing, as Mr. Michel-Levy informs me, crystallites of the cubic system. I call this constituent "Martensite," after Professor Martens, who has made in Germany, since 1878, an independent centre for metallographic researches, and has since then followed these studies with much perseverance, talent, and success' (Osmond). The term is applied solely to quenched and magnetic metal."

The outstanding features of the martensite here described are—

(a) its uniform hardness,

(b) its high magnetic permeability.

Consider last the definition of troostite:

"TROOSTITE.—(Ger. Troostit, Fr. Troostite.)

This constituent, according to Osmond, is contained in steels of

various carbon contents on quenching at certain temperatures. It is revealed by polishing the steel on parchment, moistened with a solution of liquorice root or nitrate of ammonia. It readily assumes a brown color by such treatment. It is relatively softer and more rapidly acted upon by acids than martensite. Its composition and molecular construction has not been determined.

It is easily recognizable on etching with hydrochloric acid (1 cubic centimetre) in alcohol (100 cubic centimetres). It colours dark, while martensite remains entirely unaffected in colour.

According to the newest researches on expansive properties by H. Le Chatelier, it is an intermediate formation which occurs on the transformation of martensite into pearlite and vice versa.

The definition here is somewhat vague, but two important points of difference between this constituent and martensite are referred to—

(a) the softness of the constituent relative to martensite,

(b) the readiness with which the constituent is attacked by etching reagents.

Time and time again since this glossary was compiled, the fact has been noted that samples of steel cooled at rates which are just insufficient to produce full hardness are characterized by sections which are more readily darkened on etching than others. In this connection one need only refer to the work of Portevin and Garvin, who in Fig. 34 of their paper on the Experimental Investigations of the Influence of the Rate of Cooling on the Hardness of Certain Steels, which we here reproduce, exemplify this effect very nicely. When it is remembered that only those samples which showed no darkening at their centers exhibited full hardness, the essential differences between martensite and troostite, as defined above (i.e., in terms of the glossary above referred to), are most clearly demonstrated.

In view of the marked difference in hardness between the light etching and dark etching constituents, the writer cannot conceive how they can be confused and what justification on this account can be given for referring to the needles in such a martensitic structure as is shown, for example, in Fig. 5 of the authors' paper, as troostite (?) rather than martensite. For his part the writer would prefer to retain the older definition, probably with some changes.

That both the martensite and the troostite of the glossary may be colloidal in nature appears to be of considerably less importance than the fact that they differ so widely in hardness and in their resistance to attack by etching reagents. The fact also that the martensite of freshly quenched steel may differ from that of steel aged at room temperature does not enter into the question, since the variations in the hardness of martensite with time are of an order of magnitude not to be compared with that of the difference in hardness between martensite and troostite, as defined in the glossary.

The observations we have made with regard to the resistance of steel in various states to the attack of etching reagents may raise in the minds of some the question, whether the emphasis now laid on more strictly physical investigations to the detriment of more strictly chemical

investigations of steel does not delay our advance toward a knowledge of the fundamental nature of the material. Dr. Sauveur asks on page 214 of his paper: "What direct evidence have we of the presence in hardened steel of numerous cementite particles of sub-microscopic size?" So far, one of the nearest approaches to a reasonable answer to this question appears to the writer to have been given by Whiteley in his Report on the Eggertz Test for Combined Carbon in Steel (Carnegie Scholarship Memoirs, 1917, Vol. 8, p. 1). Whiteley, by direct chemical methods, has arrived at the conclusion that: "Although the form in which the carbon exists in quenched steel has not been definitely ascertained, the evidence . . . . . indicates that it is still present as cementite," and to the same conclusion the writer was led on reading his most valuable paper.

Another conclusion one could not but reach at the same time was that still further work along this line would well be repaid. Correlation of the effect of various cooling rates on, for example, the color values (Eggertz Test) of a series of steels would be of immense interest to the physical metallurgist and would, one feels, go far toward answering the question propounded in Dr. Sauveur's paper, regarding the form of carbon in hardened steel.

While we lay stress on physical methods of investigating steels, do not let us underestimate the value of chemical methods of attack in this connection.

Written Discussion: By R. L. Dowdell, U. S. Bureau of Standards, Washington, D. C.

Dr. Sauveur has admirably brought to our attention many of the important factors relative to the structural changes taking place in various carbon steels when cooled at different rates from their respective melting points.

He has stressed a very important point in his statement, namely, "that ferrite separates from the austenite along some of the crystallographic planes giving a structure resembling the Widmanstatten pattern." This appears to be the case almost regardless of the speed of cooling the steels. Some of us are glad to know that Dr. Sauveur regards the structure of a quenched steel as consisting of martensite needles in a groundmass of austenite.

It seems that Dr. Sauveur's suggestion that the so-called martensite needles be called troostite has many points in its favor. For instance, we know that as we temper steels, containing these white martensite needles, at low temperatures, such as 100 degrees Cent., that the needles gradually become darker after etching. This, of course, shows decidedly that there is really no sharp line of demarcation between the martensite and the troostite plates or needles. On further tempering the dark needles will become lighter as either or both the time and the tempering temperature increases until we have fairly large carbide particles embedded in a ferrite groundmass. This carbide particle size can be made large enough to recognize at fairly low magnifications.

When we approach less drastic quenching, we will tend to eliminate

the white-needled formation and gradually change to the darker needles until we have the typical dark-needled troostite. If the above reasoning is accepted we might call the white martensite needles a form of spheroidal pearlite because as we approach a more stable equilibrium condition on cooling or on tempering there will be no sharp line of demarcation between martensite needles and troostite needles and also between troostite and sorbite.

When we are dealing with hardened steels, we are dealing with structures at room temperature which are in a metastable condition. Under these conditions the hardest carbon steel must contain some white martensite needles which has considerably different physical properties than the typical troostite needles or the intermediate products. Until we know definitely just what the mechanism is of austenite decomposing through martensite to troostite on quenching, it seems that the term "martensite needles" should not be dropped. Most of us probably believe that the main difference between martensite and troostite plates is the difference in the relative size of the carbide particles and ferrite grains but still we have no definite proof that physically the white needles and the dark needles are the same.

If the conditions in rapid cooling approached a more stable equilibrium and if there were no recrystallization of iron to contend with, it seems that it would be well to eliminate some of the names of our microconstituents but as it is, why not keep the controversial microstituents until we find out what they really are?

In conclusion, I wish to congratulate Dr. Sauveur and also his coworker, R. H. Aborn, for their notable contribution on this important question.

Written Discussion: By H. H. Lester, Watertown Arsenal, Watertown, Mass.

We are fortunate that so distinguished a scientist as Professor Sauveur has given us these new data and this latest expression of his views on the problem of martensite, a clear concept of which is of fundamental importance to an intelligent understanding of steel.

The picture he presents of alpha iron forming along crystallographic planes in austenite grains is a logical one and seems to give a ready explanation to the needle structure. The writer can agree with the first point, but is not quite convinced as to the needle structure. The needles are regarded as troostite and evidently are supposed to represent regions of initial formation of alpha iron. However, he neglects to explain the cored structure of these needles that has been shown very strikingly in the remarkable pictures of Lucas, some of which are shown in Professor Sauveur's book. The needles in the present view are regarded as held in a groundmass of austenite.

Some years ago the writer investigated by means of X-rays a series of carbon steels to discover among other things if there were appreciable quantities of austenite in martensite. Some gamma lines were found, but they were very faint. There appeared to be at the best less than 1 per

cent of austenite present. This amount seems to be too small to be consistent with Professor Sauveur's concept. The alpha lines were of course diffused and were displaced from normal positions indicating an enlarged alpha lattice. On tempering at successively higher temperatures up to 600 degrees Cent. (1110 degrees Fahr.) there was found to be a gradual decrease in the alpha lattice and a gradual sharpening of the spectrum lines. Carbide lines appeared after tempering at around 250 degrees Cent. (480 degrees Fahr.)

This evidence seemed to suggest that alpha iron retains appreciable amounts of carbon in solution on cooling from the quench. Some additional support to this view has been obtained in a recent study of plastic deformation where bands were found along boundaries of ferrite grains in annealed carbon steel that seemed to represent an expulsion of carbon from the grain due to the influence of cold work.

In view of the fact that there seems to be too little austenite present to form the matrix for the troostite, and in view of the evidence that carbon does remain in solution in the alpha iron, the writer has been led to regard the matrix of the needles partly as a solid solution of carbon in alpha iron rather than as austenite. It is not necessary that the groundmass should be altogether of the solid solution. No doubt it contains considerable precipitated carbide. This mixture of uncoagulated carbide in ferrite supersaturated with carbon really constitutes troostite.

Professor Sauveur's theory explains the fact, is indeed built around that fact, that the needle is harder than the matrix. He has neglected the fact of the ribs of the needles. The writer has never found the black ribs in any but hypereutectoid steel. It was pointed out to him by Mr. Lucas and later verified in various hypoeutectoid steels that alpha grain boundaries tend to form down the center line of martensitic needles, that is, there appears to be something in the needle which corresponds to a mid rib even in hypoeutectoid martensite.

Another fact of observation that is pertinent is that the needle structure tends to persist through subsequent heat treatments. For instance in some 3 per cent nickel steel studied recently by the writer the metal was quenched and tempered so as to give a sorbitic structure on first examination. After a rather deep etch followed by polishing and a light etch, the structure showed many martensitic needles. In fact the final picture might have been mistaken for martensite. Alpha grain boundaries were evident splitting the needles. More than that, the halves of the needles seemed to form separate grains although the portions of the boundaries along the midribs were much better defined than were those along the outer edges of the needles.

The facts that alpha grain boundaries follow the mid ribs of the needles and that the needle structures tend to persist as independent grains might be regarded as evidence that the needles represent regions of disturbance along atomic planes in austenite grains. Stresses set up in austenite grains due to the formation of ferrite find relief in miniature slips along planes in the austenite grains normally favorable to slip. The

path of the actual slip would be a region of maximum disturbance. It would correspond to the mid-rib of the needle. The metal adjacent to the slip would also be disturbed and would remain in a strained condition, probably breaking up into a large number of submicroscopic alpha crystals. The strained condition or the very fine grain structure in this metal of the needle would account for the relative greater hardness of the needle over that found for the matrix material.

The internal stresses causing slips would be three dimensional and the slip planes would tend to appear along conjugate octahedral planes giving the typical acicular structure characteristic of martensite. Slip planes as ordinarily observed are adjustments to one dimensional external forces. It would be expected that in the heterogeneous structure of steel occasional cases would be found where the internal stresses were predominantly in one direction and that martensite needles suggesting parallel slips would be found. The writer has found one such case.

The above concept of slips to account for needle structure necessarily brings up the question, when do these slips occur, in an early or in a late stage of martensite formation? Obviously they cannot start till stresses are developed. Hence, they must on the view expressed above represent final stages of martensite formation instead of marking initial stages as called for in Professor Sauveur's theory.

The writer feels that the initial precipitation of ferrite might and probably does occur along crystallographic planes in austenite as suggested by Professor Sauveur. He considers, however, that this initial precipitation is not responsible for the needle structure. If austenite is present as it may be especially if alloying constituents tending to stabilize austenite occur in the steel such austenite would occur in the matrix metal, but its presence would not alter the argument for the nature of the needles.

Written Discussion: By H. M. Boylston, professor of metallurgy, Case School of Applied Science, Cleveland.

Dr. Sauveur's paper gives the most complete, simple, and acceptable explanation of the hardening of steel by quenching that has ever been printed and our Society is to be congratulated for obtaining this paper for its TRANSACTIONS. Dr. Sauveur speaks with the authority of nearly forty years experience with this very problem. There have been hardening theories galore and Dr. Sauveur himself has contributed to their development not a little, but all of the previous theories fell short of complete clearness, lacked some essential feature, or failed to check with some important experimental fact or facts.

As many of our readers are aware, Dr. Sauveur has discussed in his book[1] all of the important hardening theories. Of the more recent theories the "slip interference theory" of Jeffries and Archer seems to me the most important, and undoubtedly has clarified or crystallized our ideas as to the meaning of hardness itself and the means through which ductility makes itself evident in metals, and these eminent writers have been and

[1]Albert Sauveur, "The Metallography and Heat Treatment of Iron and Steel, 1920 Edition," Chapter 18, p. 308-314, inclusive; 1926 Edition, Chapter 16, p. 251-267 inclusive.

must always be given due credit for their experimental work and theoretical deductions, but the general theory of slip interference as a cause of hardness is really more of an axiom than a theory and means just what it states and no more, i. e., that when slip in metals is interfered with, hardness results. It does not explain in what way slip is interfered with when steel is hardened by quenching. And sub-theories were promulgated by Jeffries and Archer and others, but again these sub-theories like the older theories appeared inadequate for the reasons expressed by Dr. Sauveur in his present paper.

Lucas must also be given credit for helping to elucidate the nature of martensite by the use of high power photomicrography and skilled technique in polishing, etching, and photography. But it needed Dr. Sauveur's clear logic and keen interpretation to place the matter again before us free from mystery and in accordance with the known experimental facts. Who would dare to say that we shall never know any more about hardening and tempering than is known at present! It would be foolish to make such a statement, but until new facts are brought out which refute Dr. Sauveur's arguments or answer the pertinent questions on page 214 of his paper with positive evidence, I shall accept his latest teachings with a sense of satisfaction.

Written Discussion: By Dr. W. H. Hatfield, Brown-Firth Research Laboratories, Sheffield, England.

I must regret my inability to present my contribution to the discussion in person; particularly so, as Dr. Sauveur's present valuable paper raises so many interesting questions. His critical analysis of the mechanism of the decomposition of austenite merits the most earnest consideration, and, I think, can be accepted. As to the cause of the hardness of hardened steel, I am in agreement with Dr. Sauveur, with the remarks made on pages 211 and 212, in which the hardness is suggested as being in part due to five different causes, no doubt in varying degree. I personally believe that the solid solution of carbide in gamma iron is present in the quenched carbon steel. I believe that hardened steel contains a substantial proportion of broken-down solid solution, i. e. alpha iron with extremely finely dispersed carbide: the fineness of this dispersion may be such as to approximate to the colloidal state. It may be that some carbide is retained in forced solution in the alpha iron, but that I doubt very much. It would appear to be incontestable that internal stress of high magnitude is present. We have always the fact of the very fine structure, and the effect described by Dr. Jeffries. Personally, I am prepared to leave the question in this position for the time being.

Dr. Sauveur makes the analogy with the hardening of duralumin. I would suggest that perhaps it would be more profitable to consider two other causes of hardness in the case of steels where a higher hardness is obtained without quenching, namely,

(1) The surface hardening of chromium-aluminum steel at 500 degrees Cent. (930 degrees Fahr.) in ammonia.

(2) The cold work hardening of nickel-chromium austenitic steels,

In these two instances I think it would be a fair inference that the effect of small grain size, and the dispersion of particles of carbide, are eliminated. One can understand the existence of the high internal stress, in the first case through the creation of a new chemical compound in situ, and in the latter case as the residual stresses from the cold working operations. Also one can agree to the distortion effect of the space lattice of the crystals. However, as I am one of those who are frank enough to despair, at the moment, of explaining the hardness of hardened steel, it would be drawing an even longer bow to attempt to explain the two cases which I have brought forward, and I therefore simply advance my view of the necessity of explaining all three instances at the same time.

R. F. MEHL: It is perhaps somewhat revolutionary to suggest at a meeting of American metallurgists that the precipitation-hardening theory for alloys of the duralumin type is not entirely accepted by everybody. The precipitation theory as developed for duralumin has been applied to the steels to account for the hardness of quenched austenite. Actually there is a considerable body of work, done largely in Germany by Fraenkel, Sachs and others, which shows, I think without any question, that the chief agent in the hardening of the so-called age-hardening alloys is not a precipitation. It is an effect which accompanies precipitation but which is separate from it in the sense that there are two separate effects. This is most clearly shown by the recent work of Frhr. von Göler and G. Sachs *(Metallwirtschaft,* Vol. 8, 1929, p. 671-680). Quenched duralumin, both in the polycrystalline and the monocrystalline state, aged one-half hour at about 150 degrees Cent., showed a pronounced maximum in tensile strength. The careful determination of the side of the unit face-centered cell which accompanied this work showed without question that no precipitation had occurred with the treatment. Precipitation follows later, with aging at a higher temperature, and is accompanied by an increased tensile strength not, however, so marked as in the first case. It is thought that the first hardening is caused by a preliminary agglomeration of copper and aluminum atoms into molecules preceding actual precipitation.

Professor Sauveur suggests on page 212 that the hardness of retained austenite may be conditioned by its supercooling in such a way that the austenite retained at room temperature is harder than normal austenite. I should like to suggest, then, and additionally to complicate matters, that the hardness of retained austenite may be influenced by an effect preceding actual precipitation, similar to that discovered for duralumin, and inducing an increased hardness and tensile strength.

It is difficult to imagine a more important question in physical metallurgy than that of the nature of solid-solid transitions. Professor Sauveur has continued to direct our attention to the mechanism of such transitions and rightly emphasizes the dominant role played by the crystallographic changes producing the Widmanstatten structure. It is needless to say that this present paper presents the clearest picture available of the transitions resulting from the decomposition of austenite.

Author's Closure

Dr. Honda makes me say things that I have not said. I do not con-

sider hardened steel to be made up essentially of ferrite and finely divided cementite particles. I have referred repeatedly to the likelihood of carbon being in solution in alpha iron in the needles. It will suffice to quote from page 211 of my paper where I attribute the hardness of quenched steel in part to "the hardness of the troostitic needles which probably consist of a supersaturated solid solution of carbon in alpha iron and of numerous finely divided particles of cementite." There is nothing in Dr. Honda's remarks which would justify a modification of the view I have expressed in regard to the nature of hardened steel. I consider it to be an aggregate of hard needles embedded in retained austenite.

Dr. Honda considers martensite as a solid solution of carbon in alpha iron and troostite as alpha iron containing cementite particles. It seems highly probable to me that neither of these constituents are found in quenched steel, the needles containing both carbon in solution and dispersed cementite. According to Dr. Honda, they would be neither martensite nor troostite. I would have no objection to calling martensite a solid solution of carbon in alpha iron but where is that constituent to be found?

Contrary to Dr. Honda's statement, I do not believe that during heating and cooling the total magnetic change occurs suddenly and abruptly. I have lived too long for that. I have learned from Osmond more than 25 years ago that the magnetic change, while covering a considerable range of temperature, starts with a jerk and stops with a jerk, and it is that jerk which produces the thermal disturbance known as the A_2 point, an obvious manifestation of a change of internal energy regardless of any theory of the magnetic properties of matter.

I am gratified that my paper should have called forth so important a contribution from Messrs. Portevin and Chevenard. As they deal, however, with the hardening of some alloy steels which is outside the field I attempted to cover, I shall not at this time venture to discuss it but I should like to express my high appreciation of its importance.

The kind words of Mr. Brearley are much appreciated, coming as they do from a master in clear thinking and clear expression coupled with true scientific modesty. In reply to his allusion to the aging of steel, I appreciate, I think, its importance. The thought I meant to express was that aging after quenching does not in the case of steel result in an increase of strength comparable to that observed in some nonferrous alloys, notably in duralumin. Aging does not play an important part in the hardening of steel by quenching. That is why I find it difficult to conceive the hardening of steel and the hardening of duralumin as being due to similar causes. We cannot fail to be impressed by Mr. Brearley's statement that the toughness (Izod value) of steel is "enormously lessened" by aging.

I have read with much interest and profit Professor Thompson's remarks. It is gratifying to find that he agrees with many of my statements. I shall reply to his criticisms in the order they were presented. On page 204 of my paper I stated that the alpha phase "forms first along some of the crystallographic planes of the gamma iron and around the grain boundaries." I did not say that it formed at the grain boundaries

before forming at the crystallographic planes. In expressing the belief that the alpha phase locates at the grain boundaries as well as at the crystallographic planes I was, perhaps unconsciously, proceeding by analogy—having in mind the location of ferrite (alpha phase) at the boundaries of the grains of hypoeutectoid steel. There seems to be no good reason to think that the same mechanism does not prevail in the absence of carbon. It is true, however, that the dark line in my Fig. 3 probably results from the volatilization of iron.

If the martensitic needles which I consider to be of the nature of troostite do not contain many finely divided particles of cementite, they must represent a solid solution of carbon in alpha iron. Under very high magnification, however, as shown by Lucas and others, these needles acquire a mottled appearance very suggestive of the presence of two constituents. On applying the Eggert test (color method) to the determination of the carbon in martensitic steel, it is found that about one-half of the carbon is "missing"; that is, it fails to impart any coloration to the solution, the other half being present as the carbide Fe_3C and being responsible for the resulting color. This so called "missing carbon" is supposed to represent carbon in solution. I shall venture to refer to my paper, "Notes on the Microstructure of Hardened Steel and the Missing Color Carbon," published in the *Journal* of the Iron and Steel Institute for 1896. May we not also offer as an evidence of the presence of cementite in the needles the readiness with which they are generally colored by nitric acid?

I am glad that Professor Thompson calls our attention to the possible part played by twinning in the hardening of steel. His remarks will be appreciated by many.

In regard to Professor Thompson's pertinent observation that in steel quenched during Ar_1, the pearlite formed is not found at the crystallographic planes. I should like to make it clear that I had in mind the possible formation of pearlite and retention of austenite in the same sample. Professor Thompson refers to aggregates of pearlite and martensite and thus implies a cooling relatively so slow that the pearlite, although having formed at the crystallographic plane, was given the necessary time to coagulate into "rounded patches." I believe that the mechanism of the transformation of austenite is quite independent of the amount of carbon present. It applies to eutectoid steels as well as to hypo- and hypereutectoid steels.

In reply to Professor Ellis I am glad that I brought up Professor Ellis with a jerk when I expressed the opinion that our material civilization would be in a sad plight if austenite failed to transform, since this led him to speculate so interestingly on what might be done with austenitic steel, should we be deprived of pearlitic steel. I am not going to quarrel with him on this subject, as I also believe that the possibilities of austenitic steel are far from having been exhausted. It would be idle, however, to speculate as to what the metallurgists of a nonpearlitic age might have done with austenitic steel in the absence of pearlitic steel, but I am

willing to go as far as Professor Ellis in conceiving that they would have accomplished much. It is a relief to have Professor Ellis' opinion that the American Society for Steel Treating would have come into existence after all.

In stating that all steels would remain nonmagnetic if austenite did not transform, I simply had in mind that gamma iron is nonmagnetic. It may be that metallurgists would have found a way of making it magnetic, but that would imply a transformation, the possibility of which I have excluded.

I am fully aware that committees have been appointed by various societies to define the various microscopical constituents of iron and steel. Indeed, I have taken an active part in the labor of some of them. I do not see, however, why the nomenclature they have recommended should keep me from considering martensitic needles to be of a troostitic nature.

I appreciate Professor Ellis' reference to the use of the Eggert method as a means of studying the condition of carbon in steel. It is a method that I have used in common with others some twenty years ago when it was much in vogue. I believe with Professor Ellis that it should be revived and that it is capable of yielding important results. I welcome Professor Ellis' evidence of the existence of cementite particles in hardened steel.

I welcome Dr. Dowdell's opinion that my view of the nature of the martensitic needles "has many points in its favor." Few, so far, have been willing to admit it. If the so-called martensitic needles differ from troostite, it must be (1) because the needles contain carbon in solution and troostite not, (2) because the needles contain more carbon in solution than troostite, or (3) because the needles contain carbide particles in a finer state of division. Is there any one prepared to substantiate the claim that the first assumption is true? And if it is only a question of how much carbon is in solution or of the size of the carbide particles, where is the demarcation? Will some one indicate a means of distinguishing between martensitic needles and troostite? When on tempering does a martensitic needle become troostite?

Dr. Lester holds that the so-called martensitic needles are embedded in a solid solution of carbon in alpha iron rather than in undecomposed (retained) austenite. Such a view, I believe, has never been expressed before and naturally commands our attention. The stronger argument he offers in support of it is found in his statement that there is at best less than one per cent of austenite present in quenched steel. Certainly, if he is right, so small an amount is inconsistent with my view of the nature of martensite which I regard as an aggregate of troostitic needles embedded in undecomposed austenite. From the work of other X-ray experts, however, it is permissible to believe in the presence of much larger proportions of austenite, that is, of gamma iron in quenched steel. I shall venture to refer to an article of Seljakow, Kurdumoff, and Goodzow in *Revue de Metallurgie* for April 1928 in which they show diffraction patterns of quenched steel which would indicate the presence of some 10 to

25 per cent of gamma iron. Fink and Campbell, TRANSACTIONS, American Society for Steel Treating, May 1926, refer to the presence of gamma lines intense enough to be measured. This could not be done, I believe, if the steel did not contain much more than one per cent of austenite. In his Howe lecture, American Institute of Mining and Metallurgical Engineers, April 1925, Dr. J. A. Mathews stated that gamma iron or austenite is always present with martensite in all normally hardened steel, a view which is very widely held—quoting E. C. Bain. Dr. Mathews refers to the presence of 6 to 10 per cent of gamma iron in steel containing 0.90 per cent carbon and 2 to 3 per cent chromium. Honda and Sekito[8] state that austenite is always found with martensite. From their intensity measurements it would seem that some 25 per cent gamma iron must be present in quenched steel.

Assuming the steel to be eutectoid in composition, the alpha iron forming, according to Dr. Lester, the groundmass of the martensitic aggregate would retain some 0.85 per cent carbon in solid solution which, in view of the known very limited solubility of carbon in alpha iron appears an excessive degree of supersaturation. Again the groundmass is known to be softer than the needles and this is readily explained on the assumption that it consists of austenite. Can this be accounted for, if we consider it as alpha iron holding carbon in solution? Dr. Lester refers to the midribs found, particularly by Lucas, in some martensitic needles. I have at present no personal view worth expressing in regard to their nature, but it may be well to note that they are only occasionally present.

Finally, Mr. Bain told me that in his opinion eutectoid steel, normally quenched steel, does not contain less than 10 per cent of gamma iron which is quite enough for my needs.

I find much satisfaction in Professor Boylston's acceptance, for the moment at least, of my opinion regarding the hardening of steel. Many times in the past I have discussed metallurgical questions with him and I have always derived a sense of security from his acceptance of my views. I felt, then, that they were at least reasonable and not unworthy of publication.

To have Dr. Hatfield accept so much of what I attempted to express in my paper is very gratifying and increases my confidence in the correctness of my views. Dr. Hatfield calls our attention to the surface hardness imparted to chromium-aluminum steel by treatment with ammonia at 500 degrees Cent. and to the hardness resulting from cold working nickel-chromium austenitic steel. Are these to be explained by one of the theories so far advanced to account for the hardening of steel by quenching or must we seek elsewhere for an answer? The inquiry is well worth the careful thought of those interested. It does not seem as if what we may call "nitride hardening" can be explained by the grain-size theory. Those who believe in the dispersion theory might claim that it results from the dispersion of nitride particles, while the advocates of the

[8]Kotaro Honda and Sinkiti Sekito, "On the X-ray Investigation of the Formation of Martensite," Science Reports, Tohoku Imperial University, Sendai, Japan, Vol. 17, No. 4, 1928, p 743

distortion theory might reasonably argue that the introduction of nitrides within the space lattice causes distortion, hence increased hardness. As for the hardness of cold-worked nickel-chromium austenitic steel, it seems to result from a superficial and partial decomposition of austenite into a martensitic aggregate, the hardness of which may be explained in accordance with our view of its nature.

Professor Thompson has called attention to the intersection (approximately at the melting point of iron) of the density-temperature curves for the face-centered and body-centered lattice phases. At the melting point of iron, it seems logical to expect that at the moment of its formation any stable solid phase (necessarily in equilibrium with the liquid phase) should have equivalent properties with any other stable solid phase. Or, in other words, they all start from "scratch," but immediately diverge with an infinitesimal decrease in temperature. It is, of course, true that with one component there can actually be but one solid phase in stable equilibrium with the liquid phase. But if at the moment of formation two solid phases, such as face and body-centered iron, had equivalent properties they would in effect be but one phase. Thus, if it were possible to preserve gamma iron as a stable phase beyond the A_4 point, its density-temperature curve should intersect that of delta (alpha) iron at the melting point.

Experimentally observed data on the density of liquid iron are admittedly inexact and of a preliminary character. Berlin, whose work Professor Thompson regards as most reliable, quotes his results as tentative and discusses several "factors which may exert an erroneous influence on the determinations." The somewhat earlier work of Benedicks, Berlin, and Phragmen[9] indicated a density of 6.92 ± 0.07 of iron just above the melting point, "in good agreement," as Desch and Smith state[10] with their own recently determined values. It is true that the X-ray data may be inexact, due to the extreme difficulties of high temperature examination and undoubtedly need confirmation by a larger number of investigators. Yet, it seems to be a case in which the pot can scarcely call the kettle black.

Professor Honda, I believe, is incorrect in stating that Sato from X-ray data calculated volume changes of 0.0082 per cent and 0.0023 per cent, respectively, for the A_4 and A_3 transformations. In the reference cited,[11] Sato calculated the linear changes and reported 0.08 per cent and 0.27 per cent in agreement with his observed values. The volume change bears no simple relation to the linear change but to a first approximation may be regarded as at least three times greater.

Furthermore, Sato refers to the X-ray data used in his calculations

[9] C. Benedicks, D. W. Berlin and G. Phragmen, "A Method for the Determination of the Specific Gravity of Liquid Iron and Other Metals of High Melting Point," *Carnegie Scholarship Memoirs*, Vol. XIII, 1924, p. 129.

[10] C. H. Desch and B. S. Smith, "Interim Report on the Density of Molten Steel," *Journal*, Iron and Steel Institute, Vol. CXIX, 1929, p. 358.

[11] S. Sato, "Dilatometric Investigation of the A_3 and A_4 Transformation in Pure Iron, Science Reports, Tohoku Imperial University, Sendai, Japan, Vol. XIV, No. 5, 1925, p. 513.

as those of Westgren. The values given for gamma iron at 1000 degrees Cent. (3.65Å.) and at 1100 degrees Cent. (3.66Å.) I have been unable to find in any of Westgren's published data. He does, however, in his first (preliminary) paper to the Iron and Steel Institute (1921) present data indicating a value of 3.60Å. at a temperature of "about 1000 degrees Cent." In later work, with more accurate apparatus, the value 3.63Å. at 1100 degrees Cent. is given.[12] Wever's[13] data give 3.60Å. at 900 degrees Cent. I believe these are the only published X-ray data on the lattice constants of iron at elevated temperatures. It thus appears that the basis of Sato's calculations on gamma iron are inaccurate.

Dr. Mehl's remarks appear to me very significant. If the major part of the increased hardness of duralumin on aging precedes the precipitation of $CuAl_2$ particles, then it can hardly be claimed that such precipitation is the sole cause of the increase or even the chief cause, and the precipitation theory of hardening should be reconsidered.

May not the increased hardness of duralumin on aging after quenching be due to lattice distortion resulting first from the agglomeration of copper and aluminum atoms into molecules and later from the precipitation of $CuAl_2$ particles.

[12]A. Westgren, "X-Ray Studies on the Crystal Structure of Steel," *Journal*, Iron and Steel Institute, Vol. 105, 1922.

[13]Franz Wever, "Zur Allotropie des Eisens," *Stahl und Eisen*, Vol. 45, 1925, p. 1208.

APPLICATION OF X-RAYS TO THE STUDY OF PLASTIC DEFORMATION IN STEEL WIRES AND MOLYBDENUM RIBBON

By Norman Philip Goss

Abstract

The first portion of this paper describes the technique involved in making X-ray spectrograms.

The relative advantages of Laue and Hull patterns are discussed, and the equation $N\lambda = 2\,d \sin \Theta$ *developed by Sir W. Bragg is explained.*

The following problems were studied:

(a) The effect of slight plastic deformation upon the structure of low carbon steel wires.

(b) The nature of "radial asterism."

(c) The effect of annealing upon the cold-rolled structure of molybdenum ribbon.

This investigation showed that in the early stages of grain fragmentation no orientation of the crystalline debris takes place until the grains are all broken into fine particles of submicroscopic dimensions.

Two types of "radial asterism" were found, one due to the reflection of X-rays from curved crystalline surfaces, and the other from crystalline fragments, all of which lie in the same general direction but whose faces differ slightly in orientation from fragment to fragment.

Annealing a drastically cold-worked molybdenum ribbon causes complete recrystallization but directional properties are still present even after weeks of heat treatment.

Introduction

Historical Sketch of Previous Work

DURING the past ten years great progress has been made by using X-rays as an instrument of investigation with which to study the structure of metals, after they have received certain heat or mechanical treatments. In many instances this method of investi-

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gation when applied to metallurgical problems has revealed information which could not be studied in any other way. An example is the controversy which existed between metallurgists over the nature of beta iron. One group believed the structure of beta iron to be entirely different from that of alpha iron. They based their claim upon the established fact that beta iron is nonmagnetic while alpha iron is strongly magnetic. Another group believed beta and alpha iron to have the same crystal structure, since it is known that no volume change takes place when passing from beta to alpha iron. A volume change does take place when passing from alpha to gamma iron. Dr. Arne Westgren and Gosta Phragmen[1] using the X-ray spectrograph showed that beta iron and alpha iron possess the same crystal structure, the body-centered cubic lattice.

In 1920 S. Nishikawa and G. Asahara[2] made a study of metals by means of X-rays. They studied the effect of rolling upon aluminum, cadmium, copper, lead, silver and tin. Instead of using monochromatic X-rays as in the Hull[3] method, they used the "continuous" radiation. A Coolidge tube having a tungsten target was employed and a potential of 60,000 volts was impressed across the electrodes of the tube. This voltage does not excite the "characteristic" radiation of tungsten. From the nature of the X-ray spectrograms obtained, these investigators concluded that a rolled structure produces ill-defined Laue patterns, with a symmetry related to the direction of rolling.

A large amount of work has already been published[4] describing what takes place when a metal is plastically deformed as in drawing a wire through a die so as to reduce its area as much as possible without breaking it. They described what takes place when a metal was severely worked, and showed that the grain fragments orient themselves in such a way so as to have certain crystallographic planes within the grain fragments, set themselves in the direction of work-

[1] A. Westgren and G. Phragmen, "The Crystal Structure of Iron and Steel," *Zeitschrift fur Physikalische Chemie*, 102, 1922, 1.

[2] S. Nishikawa and G. Asahara, "Some Studies of Metals by Means of X-Rays," *Physical Review*, American Physical Society, 15,38.

[3] A. W. Hull, "A New Method of X-Ray Crystal Analysis," *Physical Review*, American Physical Society, 10, 1917, 661.

[4] E. C. Bain and Zay Jeffries, "Mixed Orientation Developed in Crystals of Ductile Metals on Plastic Deformation," *Chemical and Metallurgical Engineering*, Vol. 25, No. 17, October 26, 1921, p. 775-777.

H. Mark, M. Polyani, and E. Schmid, "Vorgänge bei der Dehnung von Zinkkristallen," *Zeitschrift fur Physik*, Vol. 12, 1922, p. 58.

ing. In the case of metals crystallizing with the body-centered cubic space lattice, the (1 1 0) planes of the grain fragments tend to set themselves along the direction of tension. Fig. 1 illustrates how the (1 1 0) planes of a body-centered cubic space lattice are arranged in a wire drawn through a die.

Purpose of Investigation

This paper will be devoted to the study of the following problems:

(a) The effect of slight plastic deformation upon the structure of steel wires

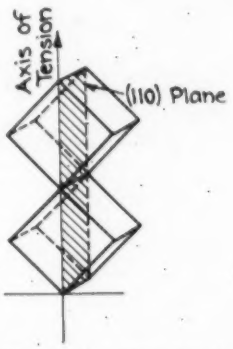


Fig. 1—Illustrating How the (1 1 0) Planes of a Body-Centered Cubic Lattice are Arranged in a Wire Drawn Through a Die.

(b) The nature of "radial asterism"

(c) The effect of annealing upon the structure of severely rolled molybdenum.

To study these problems by means of X-rays, it was found that the Hull method was not sensitive enough when studying the effect of slight deformation upon iron wires. The regular Hull method is more useful when the plastic deformation is drastic, because the diffraction lines broaden with increasing plastic deformation. By experiment it was found that the "continuous" radiation is more sensitive to slight differences in grain fragmentation produced by cold work. It is possible to detect the difference between a wire drawn 2 per cent and one drawn 5 per cent, as is shown in Figs. 16 and 17. The method of X-ray analysis used in this investigation combines the Hull and the Laue[5] methods, since each method has its distinct

[5]Max Laue, "Eine Quantitive Prüfung der Theorie für die Interferenz Erscheinungen bei Rontgenstrahlen." Sitzungsberichte der mathematische-physikalische Classe der Konglishen Bayerischen Akademie der Wissenschaften, Vol. 45, 1912, p. 363.

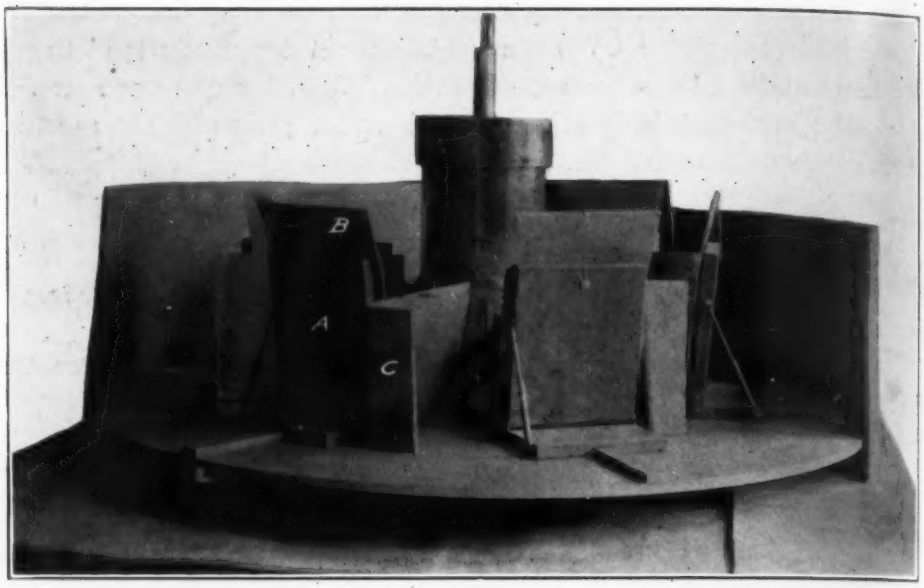


Fig. 2—Spectrograph Used in this Investigation. A—Photographic Plate Holder. B—Photographic Plate. C—Scattering Shields.

advantages. The Laue pattern gives the most information when the drafting of the wire does not exceed 15 per cent, and the Hull spectrum is more useful when the drafting exceeds 15 per cent.

To study the effect of annealing upon the structure of severely drawn or rolled wires, it is optional whether the Hull or Laue method is used. The Laue method is the more rapid and the results are the same.

The Laue method is the most useful when the nature of "radial asterism" is studied, since more detail is obtained when heterogeneous radiation is used.

Apparatus and Method of Procedure

Type of Spectrograph

Fig. 2 shows the X-ray spectrograph used in this investigation, and Figs. 3 and 4 show the types of spectrograms obtained. Fig. 3 is a regular Hull pattern produced when the X-ray radiation is monochromatic. Fig. 4 is a spectrogram having the regular Hull pattern superimposed upon the Laue.

Description of Apparatus

The X-ray tube is surrounded by a lead cylinder 3/8 inch thick,

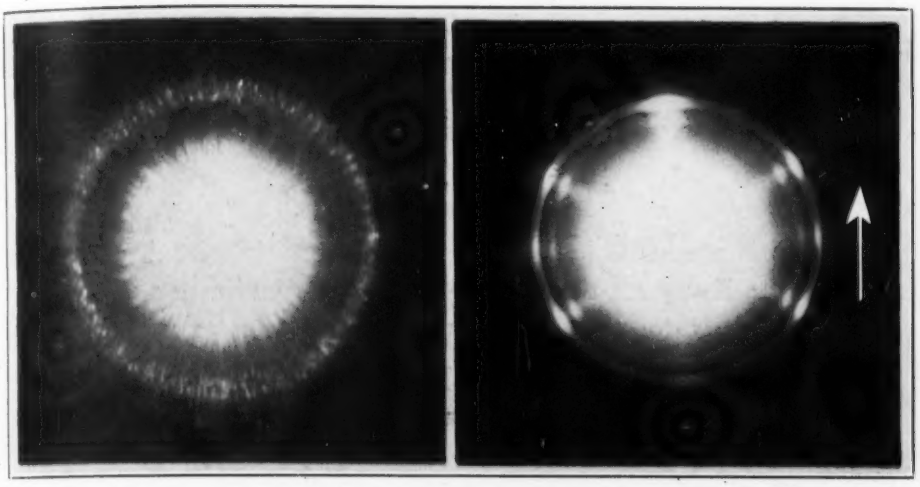


Fig. 3—Monochromatic Pin Hole Pattern of an Iron Wire. The First Ring is Due to the (1 1 0) Planes, the Second is Due to the (1 0 0) Planes.
Fig. 4—X-ray Spectrogram of a Hard Drawn Wire, Employing "Continuous" and "Characteristic" Radiation. Arrow Indicates Direction of Drafting.

which thickness is sufficient to completely absorb the X-rays. This cylinder is provided with six equally spaced orifices, so placed that the X-ray beam which leaves the target at an angle of about 5 degrees passes through them, as it is known that at this angle the beam is most intense. To prevent the high potential from breaking down between the lead cylinder and the X-ray tube, bakelite tubing able to withstand 80,000 volts, is placed between them.

The filament of the X-ray tube is heated by a current from a 12-volt storage battery and is regulated by a variable resistance connected in series with it. The filament may also be heated by means of a small step-down transformer but longer filament life is obtained when a storage battery is used, and also better control of the current through the tube.

Slit System—Advantages

The X-ray beam is defined by diaphragms with pin holes in them. Two of these lead diaphragms are placed in multiple so as to make the X-ray beam as nearly parallel as possible. This multiple arrangement of diaphragms with pin holes in them constitutes a slit system. One advantage of this particular X-ray spectrograph over any other type is the closeness of approach of the specimen to the target of the X-ray tube. The first lead diaphragm is 2¼ inches away from the target, and the specimen is placed only 5¼ inches away from the target. This makes it possible to greatly decrease the time

of exposure. With this arrangement of slits and X-ray tube it is possible to obtain an intense X-ray spectrogram of a steel sample in about eight hours, using 15 milliamperes through the tube, and impressing 33,000 volts across the tube.

Circuit and Transformer

Fig. 5 is a schematic diagram of the electrical circuit used. The target of the X-ray tube and one end of the transformer secondary is earthed. This is done because it is necessary to water cool the target,

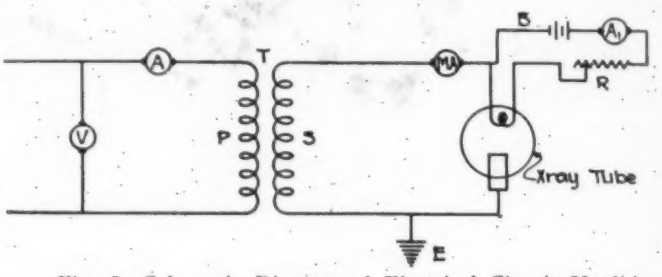


Fig. 5—Schematic Diagram of Electrical Circuit Used in this Work.

and it would be difficult to insulate a stream of water at a high potential. The voltage applied to the X-ray tube is supplied by a General Electric dental X-ray transformer, capable of producing a peak voltage of 45,000 volts.

Preparation and Mounting of Samples

The specimens for investigation must be of such thickness that the X-ray beam can pass through with maximum reflection and minimum absorption. If the specimen is too thick the X-ray beam will be absorbed without reflection, and if too thin the reflection will be too small. With experience one can readily tell how thick the specimen should be in order to have the optimum conditions. In the case of molybdenum and iron best results are obtained when the specimen is 0.004 inch thick. If the specimens are ground down on a water-cooled emery wheel they must be etched in a suitable etching solution so that all the crystalline debris formed during grinding is removed. Great care must be taken when grinding down specimens on a water-cooled wheel, since excessive grinding will introduce strains in the samples.

The specimen is mounted upon the diaphragm of the slit with shellac. In determining the effects of cold work on the grain struc-

ture it is desirable to mount all the samples so that the direction of working is always vertical. This procedure will assure uniform X-ray patterns with the working direction always known.

The photographic plate is usually placed about 5 to 7 centimeters from the specimen. Regular 4 x 5 inch X-ray duplitized Eastman (contrast) plates are used. To prevent fogging of the plates, due to scattered X-rays, lead shields are placed so as to cut off all undesirable scattering.

Discussion

Types of Lattice in Iron

The atoms within a crystal are arranged upon a space lattice. The most simple type of space lattice known is the cubic. In this

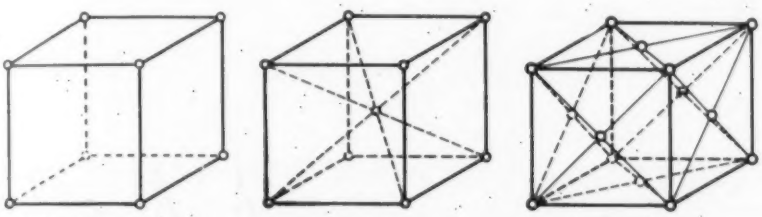


Fig. 6 Fig. 7 Fig. 8

Fig. 6—Diagram of Cubic Space Lattice.
Fig. 7—Diagram of Body-centered Cubic Lattice.
Fig. 8—Diagram of Face-centered Cubic Lattice.

type of lattice it is supposed that an atom is located on each corner of the cube as shown in Fig. 6.

However, few substances are known which have a space lattice as simple as the one just described. Many of the metals do possess a lattice of the cubic type, but modified in this way. When the cube has an atom at the center of the cube and one at each corner the space lattice is called the body-centered cubic lattice. Metals which have a lattice of this kind are alpha iron, molybdenum, copper, vanadium, etc. Fig. 7 illustrates this type of space lattice.

When the cube has an atom at the center of each face and one at each corner, the lattice is called face-centered. Metals which possess this type of lattice are gamma iron, aluminum, copper, platinum, etc. Fig. 8 illustrates the lattice of this type.

Notation Used

The planes which can be passed through a space lattice are

designated by the Miller system of indices in which the letters (h, k, L) are the reciprocals of the intercepts upon the x, y, z, axis. To illustrate, let a number of important planes be passed through the cube shown in Fig. 9.

The (1 1 1) plane cuts the x axis at 1, the y axis at 1, and the z axis at 1. A C G is a (1 1 1) plane. The 2 1 1 plane cuts the x

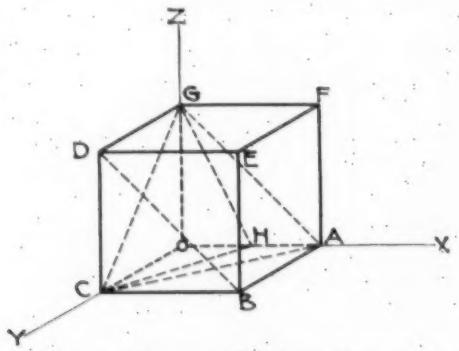


Fig. 9—Diagram Showing Manner in Which Planes May Cut a Cube. A C G is a (1 1 1) Plane. H C G is a (2 1 1) Plane.

axis at ½, the y axis at 1, and the z axis at 1. The reciprocal of these intercepts describes the plane in terms of the Miller system of indices. H C G is a (2 1 1) plane. Sometimes it is more convenient to represent a plane by its normal, thus giving the planes direction. A plane's direction is indicated when a [ ] is placed about the Miller indices, instead of the regular parenthesis ( ) as:

(1 1 0) designates a plane, and

[1 1 0] designates the plane's direction.

It has been found that the family of planes which have the largest spacing reflect the X-rays incident upon them with the greatest intensity. These planes also have the highest atomic concentration. To illustrate, alpha iron has a body-centered cubic lattice, like the one shown in Fig. 7. The (1 1 0) planes have the greatest spacing and atomic concentration. The (1 0 0) planes have the next greatest spacing and atomic concentration, etc.

How X-Rays are Reflected

When a beam of X-rays is passed through a specimen of iron, reflection takes place from the atomic planes described above. The manner in which the incident X-rays are scattered depends upon the nature of the substance. If the scatterer is amorphous the X-rays

would be scattered in all directions, but when the scattering material is made up of many minute crystals, then the X-ray beam incident upon it is scattered in definite directions. According to Bragg's equation reflection from the atomic planes can only take place when the relation $N\lambda = 2\ d \sin \Theta$ is satisfied.

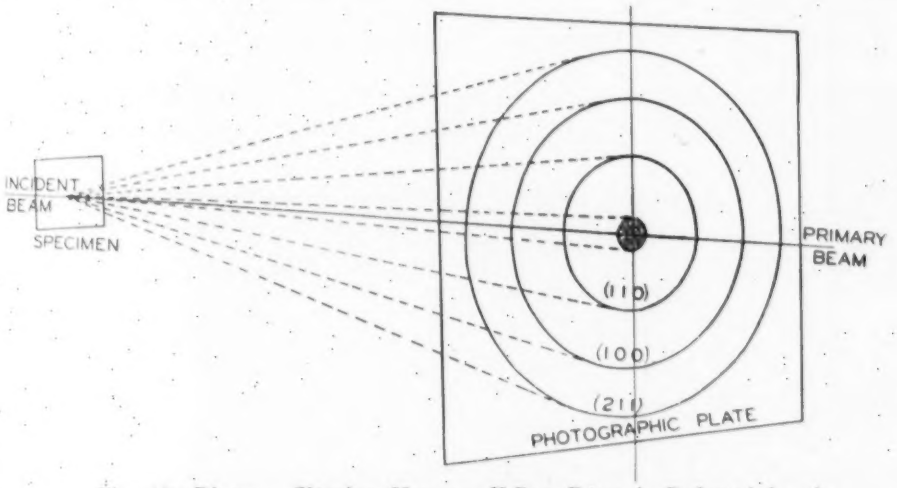


Fig. 10—Diagram Showing How an X-Ray Beam is Reflected by the Various Planes When Transmitted Through Alpha Iron.

Where

λ=wave length of X-rays.

Θ=angle the primary X-ray beam must make with crystal face for maximum reflection.

d=spacing between successive parallel atomic planes.

When an X-ray beam is transmitted through a thin specimen of iron, it is assumed that this sample is composed of small crystals oriented at random. Surely within such an iron sample there will be a great many planes set at the correct angle Θ, so that reflection will take place. Due to the mixed orientation of the planes within the sample, all the beams which are reflected at the angle will lie upon the surface of a cone, of which the primary X-ray beam is the axis. The family of planes which have the greatest spacing will reflect the incident X-ray beam through the smallest angle. Fig. 10 is a diagram showing how the X-ray beam is reflected by the various planes, when a beam of X-rays is transmitted through an alpha iron specimen. The (1 1 0) planes have the greatest spacing, thus having the smallest angle of reflection.

The reflections which occur within the iron sample are recorded

by means of a photographic plate placed at right angles to the primary beam and 7 centimeters in front of the specimen. Fig. 3 is an X-ray pattern produced by this method.

Types of Radiation Used

There are two types of X-ray radiation produced when a cathode ray strikes the target of the X-ray tube. The "characteristic" radiation and the "continuous" radiation. The "characteristic" radiation is a line spectrum, emanating from the target when properly excited.

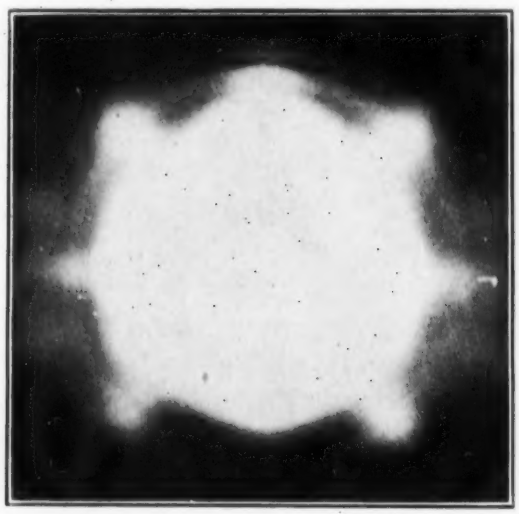


Fig. 11—X-Ray Spectrogram Showing Reflection of the "Continuous" Radiation from the (1 0 0) and (2 1 1) Planes. The Pattern is One Characteristic of Hard-rolled Low Carbon Wire.

The "continuous" radiation is a band spectrum produced when the electron stream is stopped by the target. The shortest wave length produced depends only upon the voltage applied to the tube.

The lines in the "characteristic" spectrum are called K, L, M, etc. The K lines are the hardest or shortest wave length. The K radiation consists of three lines, namely the $K\alpha$, $K\beta$, and $K\gamma$. The $K\gamma$ line is the hardest and the $K\alpha$ is the softest. The $K\alpha$ line consists of two lines very close together. The molybdenum $K\alpha$ doublet as it is called consists of two wave lengths 0.712×10^{-8} centimeters and 0.708×10^{-8} centimeters.

These radiations cannot be produced independently of one another but appear at the same time. The K radiation of the line

spectrum is produced with great intensity, and is therefore the most important radiation.

To study the effects of mechanical work and heat treatment with the X-ray spectrograph, it is not necessary to filter the X-ray radiation. As already explained above the heterogeneous X-ray radiation is the most useful. In some spectrograms the pattern due to the "characteristic" radiation can be distinguished from the one due to the continuous radiation. In Fig. 4 the patterns due to each kind of radiation can be separated with ease. The six spots which immediately surround the central spot are due to the "continuous" radiation. The first two circles close together are due to the K radiation. The inner ring is due to the Kβ and Kγ radiation, while the outer one is due to the Kα radiation. The circles and the six spots are both due to reflection from the (1 1 0) planes. The next two rings are due to reflections from the (1 0 0) planes, and the next two were produced by reflections from the (2 1 1) planes.

The "continuous" radiation is only reflected intensely from the (1 1 0) planes. Under certain conditions it is also reflected from the (1 0 0) planes and (2 1 1) planes. Fig. 11 is an X-ray pattern which shows that under certain conditions the "continuous" radiation is reflected with great intensity from the (1 0 0) and (2 1 1) planes.

Plastic Deformation

In order to explain the mechanism of preferred orientation, Polanyi, Mark and Schmid have worked out a useful method to represent in a simple manner the orientation of the planes in a drawn wire.

A sphere is drawn as shown in Fig. 12. The wire axis is taken in the vertical direction. A plane is represented by its normal. The point of intersection of the plane's directions with the surface of the sphere gives the orientation of the plane within it. So every point on the surface of the sphere represents the orientation of a plane within it, and therefore within the wire specimen which we may consider placed at the center of this reference sphere.

A beam of X-rays is now passed through the specimen as shown in Fig. 12. When the relation $\eta \lambda = 2 d \sin \Theta$ is obeyed, X-rays will be reflected from the various planes. The locus of all such planes are represented by the verticle circle E F on the reference sphere, when λ and d are known.

The two horizontal circles AB and CD represent the possible orientations in which the (1 1 0) planes will reflect X-rays of wave length λ. All the (1 1 0) planes whose normals [1 1 0] lie upon these circles belong to grain fragments which possess directional properties.

Upon examining Fig. 4 it is seen that the Kα radiation reflected by the (1 1 0) planes consists of six distinct arcs instead of a con-

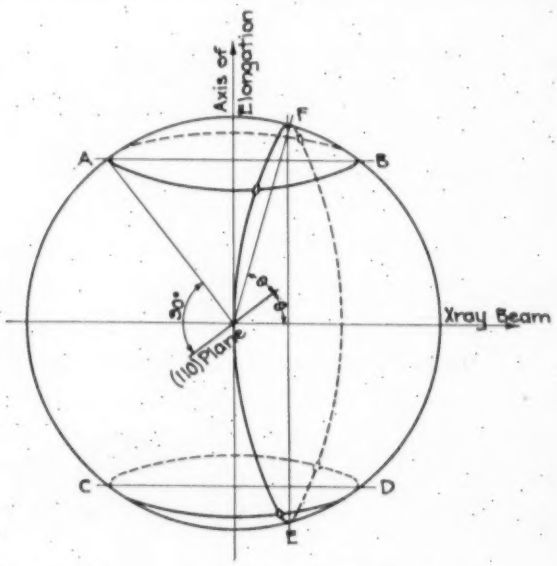


Fig. 12—Diagram to Explain the Mechanism of Preferred Orientation of Planes in Drawn Wire.

tinuous ring of equal intensity. This means that certain orientations within the wire specimen are missing. Since this ring is due to reflection from (1 1 0) planes, it is evident that the (1 1 0) planes must be definitely oriented.

If the arrow in Fig. 4 represents the direction of the wire axis, referring to the sphere it is evident that the spots which are vertical, (the two spots which lie in the direction of the arrow) are due to reflections from planes which are nearly at right angles to the wire axis, as shown in Fig. 13. To be exact, these planes are $(90° - \Theta)$ to the wire axis. In this case the angle Θ is about 10 degrees. The other four spots are due to the other (1 1 0) planes which cut the cube as shown in Fig. 14.

These planes orient themselves so that the bisector of the angle between them coincides with the axis of elongation or tension. These two (1 1 0) planes are the ones upon which slipping takes place.

When an iron wire is severely drawn through a die, the (1 1 0)

planes tend to set themselves at right angles to the wire axis, and the cubes are perfectly at random around the wire axis. This phenomena is called preferred orientation. When the (1 1 0) planes are at right angles to the wire axis another set within the cube becomes fixed automatically so that the axis of tension bisects them.

What takes place when a wire is only slightly plastically deformed and the amount of drafting necessary to introduce preferred orientation will now be considered.

Before any work could be started on this problem it was necessary to test the wire for perfect annealing. Fig. 15 is an X-ray

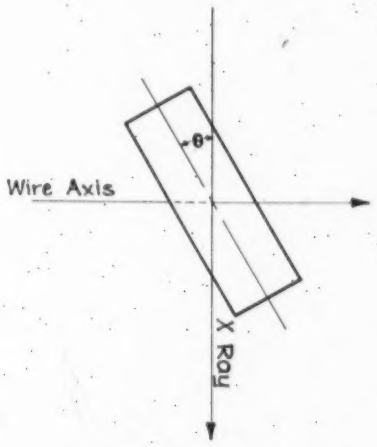


Fig. 13—Diagram Showing that Reflections are Due to Planes which are Nearly at Right Angles to the Wire Axis. θ is about 10 Degrees.

spectrogram of the wire used in this investigation. This spectrogram is only one of the twelve taken, all of which gave a pattern similar to the one shown. The target images are sharp, and those due to the $K\alpha$ (doublet) are resolved. The $K\alpha$ doublet is only resolved when the crystals are nearly perfect. The grain distribution is random.

The specimen for X-ray analysis was taken parallel to the wire axis and ground down until it was 0.006-inch thick, and etched in a 10 per cent solution of HNO_3 in alcohol. The wire from which this X-ray spectrogram was made was drafted 2 per cent. The diameter of the annealed wire was 0.375-inch in diameter. The drafting reduced its diameter from 0.375 to 0.372-inch. Fig. 16 is the X-ray pattern obtained.

Note how diffuse the target images are. The halo surrounding the central image is now drawn out. This halo is due to the "con-

tinuous" radiation and is much more sensitive to a small amount of cold work than the "characteristic" radiation. In Fig. 15 the images due to the individual grains could be seen easily, but in Fig. 16 they are completely diffuse and drawn out radially. This means that a draft of 2 per cent caused marked grain fragmentation, but no preferred orientation is present. The grains although fragmented still have random orientation.

When this wire was drafted 5 per cent the grain fragmentation is still more pronounced. Fig. 17 shows the pattern obtained. The

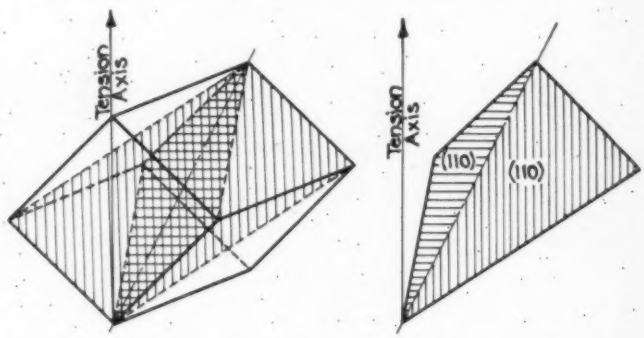


Fig. 14—Diagram showing that the Four Spots of Fig. 4 are Due to the X-Ray Reflections of the (1 1 0) Planes which Cut the Cube.

target images due to the Kα and Kβ radiation are now very diffuse. The halo surrounding the central spot, and due to the continuous radiation is now intensely radial. Preferred orientation is not evident at this stage.

If Figs. 16 and 17 are now compared it is at once apparent that it would be hard to tell which wire was drawn 2 per cent and which 5 per cent, from the Kα and Kβ radiation. It is easy to tell which has been drafted 2 and 5 per cent when the continuous radiation pattern is studied.

The wire was now drafted 10 per cent and an X-ray spectrogram made. Fig. 18 is the pattern obtained, and shows that the grains are now completely fragmented, as no target images due to the crystal faces can be seen. There are now six distinct spots surrounding the central image. This means that some of the grain fragments are beginning to orient themselves, as described above. Preferred orientation is not marked at this stage of plastic deformation.

What occurs when the wire is drafted 15 per cent is shown in Fig. 19. More and more grain fragments have oriented themselves so as to have the (1 1 0) planes lie along the wire axis. Should the

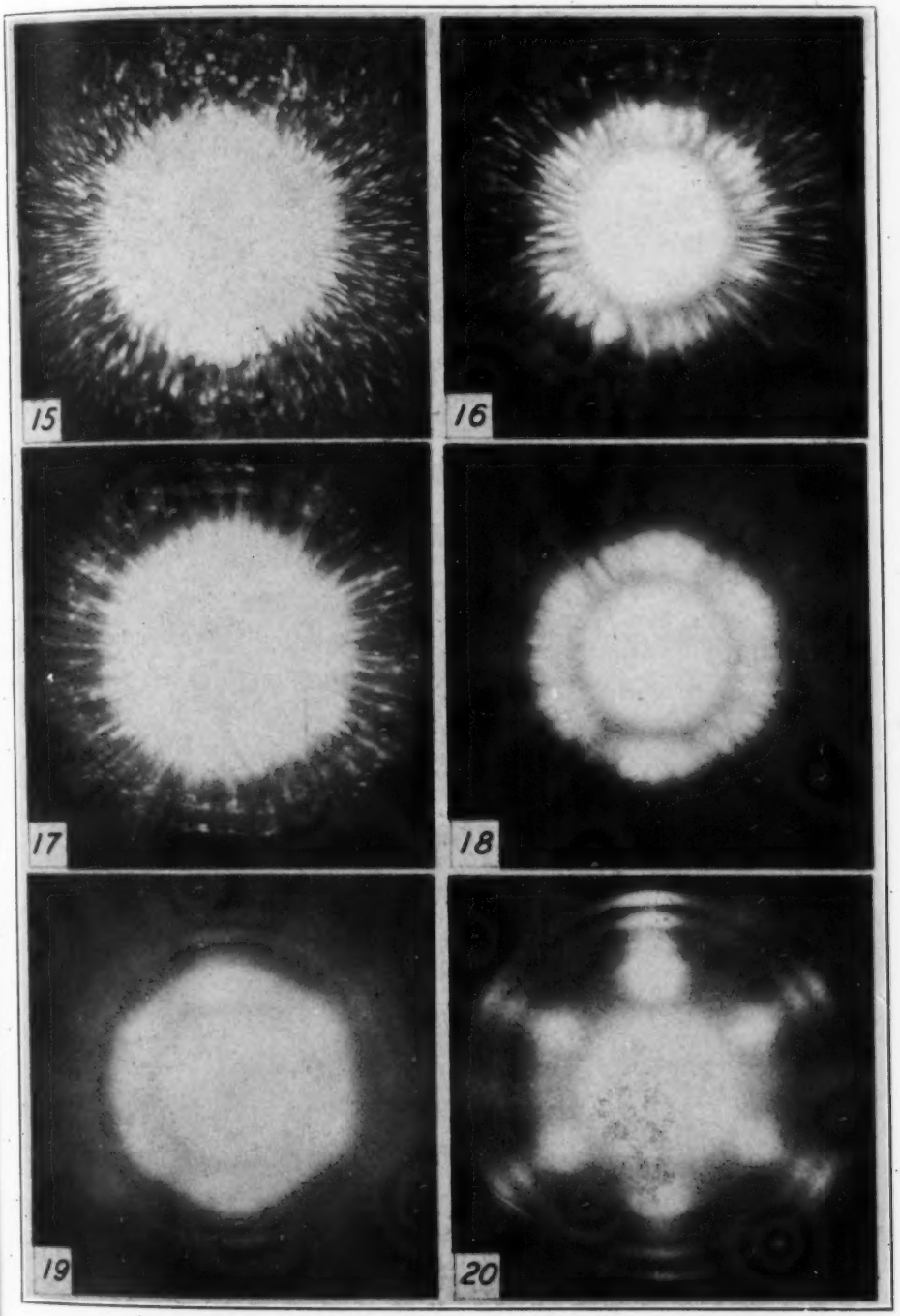


Fig. 15—Laue and Hull Pattern of a Well Annealed Low Carbon Steel Wire.
Fig. 16—X-Ray Pattern of a Low Carbon Steel Wire Drafted 2 per cent.
Fig. 17—X-Ray Spectrogram of a Low Carbon Steel Wire Drafted 5 per cent.
Fig. 18—X-Ray Spectrogram of a Low Carbon Steel Wire Drafted 10 per cent.
Fig. 19—X-Ray Pattern of a Low Carbon Steel Wire Drafted 15 per cent.
Fig. 20—X-Ray Spectrogram of a Low Carbon Steel Wire Drafted 80 per cent.

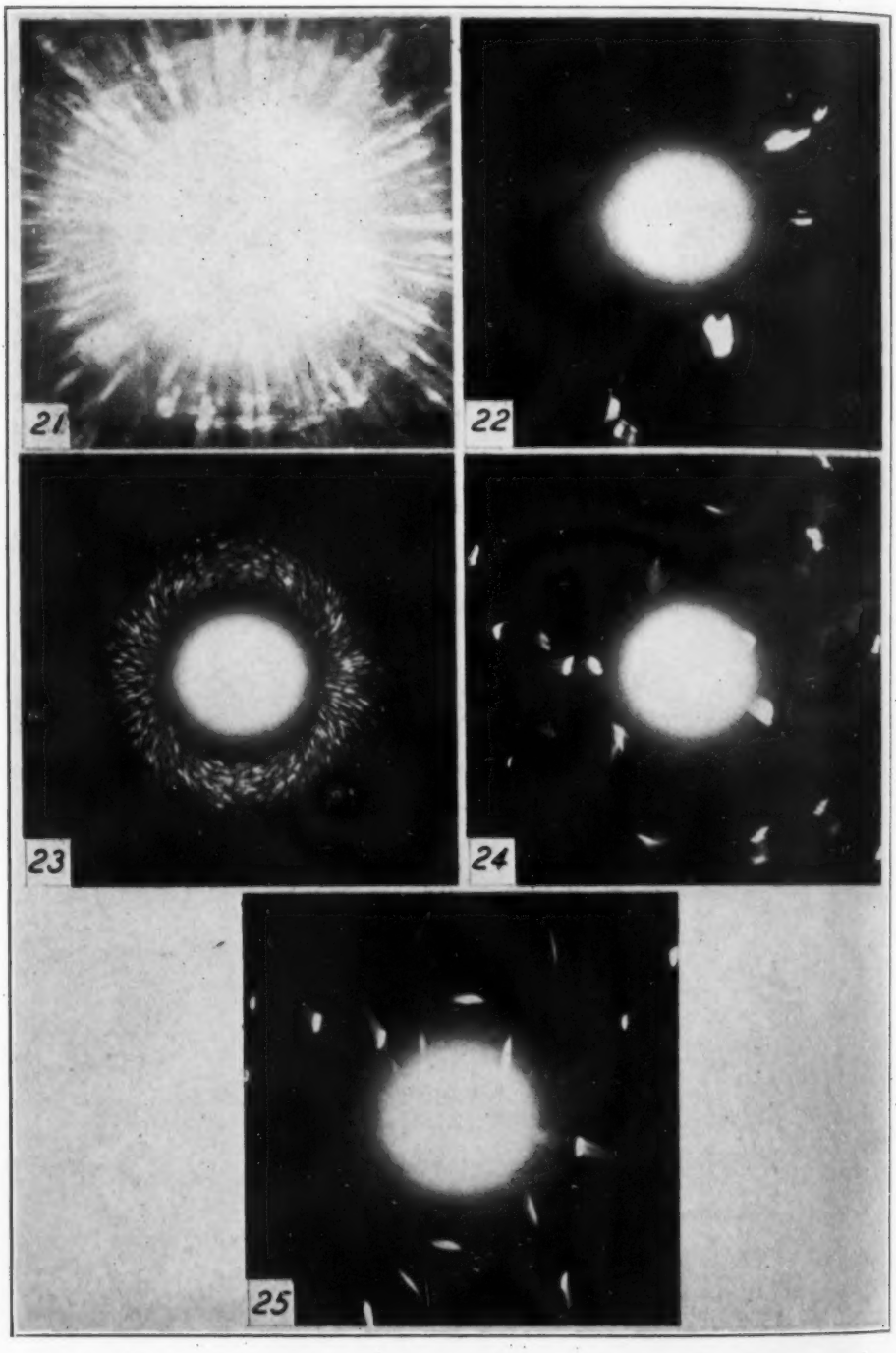


Fig. 21—X-Ray Spectrogram of Annealed Wire Showing Radial Asterism.
Fig. 22—X-Ray Spectrogram of Large Grained Armco Iron.
Fig. 23—X-Ray Spectrogram of a Wire having Small Grains.
Fig. 24—Laue Pattern of a Gypsum Crystal. Fig. 25—Laue Pattern of a Bent Gypsum Crystal.

drafting of the wire be continued more and more grain fragments will set themselves as described above. Fig. 20 is the X-ray pattern of the wire when it is drafted 80 per cent. The preferred orientation is now very intense.

From a study of large grained steel wires prepared in various ways, it was found that the wires which were brittle always gave X-ray spectrograms which showed a drawn out condition of the target images. This is illustrated in Fig. 21. It shows that the grains are large, and that most of them are in a strained condition. The lack of random grain orientation is also evident.

It appears that when steel wires are annealed so as to produce large grains, the atoms arrange themselves upon curved planes instead of flat planes. The arrangement of the atoms upon curved atomic planes introduces brittleness.

A study of large grained Armco iron wires was made by Lawrence Ott at Western Reserve University. The grains were extremely large, but the wires were extremely ductile. X-ray spectrograms were made of some of these wires, and it was found that the target images were free of "radial asterism." That this is true is shown in Fig. 22. It was also found that the more brittle a given wire was, the greater the number of grains composed of atoms lying upon curved surfaces.

This experiment shows us that whenever the grains in a wire are large and have the atoms lying upon curved surfaces instead of flat ones, the wire will be brittle. The degree of brittleness is governed by the amount of curvature and by the number of grains distorted. If the wires are composed of large grains, which have the atoms arranged upon flat atomic surfaces then the wires will be ductile. In order to produce large-grained iron wires which are ductile, it is only a question of preparing grain structure free of distortion, i. e., the atoms must lie upon flat surfaces instead of curved surfaces.

That such large grained materials can be produced is illustrated by the method developed by Mr. Ott. Some of the wire specimens examined had grains as large as ¼ inch in length yet these wires were extremely ductile.

When Figs. 21 and 22 are compared the reason is apparent, and they disprove the theory that large-grained steel materials are necessarily brittle, it is only a question whether or not the grains are composed of atomic layers lying upon flat surfaces.

An Explanation of "Radial Asterism"

A metal may be considered as annealed when the X-ray spectrogram has the following characteristics.

(1) The target images reflected from the crystal planes are sharp.

(2) The target images must be perfectly at random.

(3) When the metal is made up of nearly perfect crystals the Kα doublet is resolved.

Fig. 15 is an X-ray spectrogram of a well-annealed low carbon steel wire, and has all the characteristics just described. When the focal spots or target images are found to be diffuse then the crystals

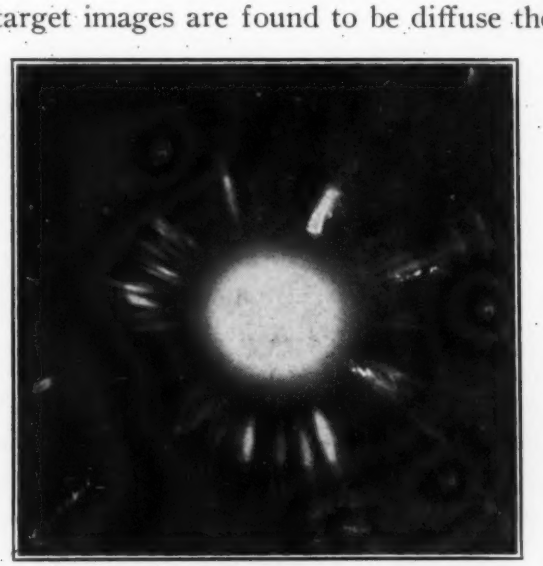


Fig. 26—Laue Pattern of Titanium Nitrocyanide.

which reflected them must be fragmented. This condition is illustrated in Fig. 17. Since the target images are diffuse, it is concluded that the grains which reflected the incident X-ray beam are fragmented.

When a wire which is composed of small grains is X-rayed, the pattern shows sharp target images. Such a pattern is shown in Fig. 23. From the nature of the pattern obtained it is evident that the atoms lie upon flat planes. To prove the statement that "radial asterism" is due to the atoms lying upon cylindrical surfaces instead of flat ones, a small slip of a gypsum crystal was obtained. Fig. 24 is the Laue pattern of such a crystal, which is already slightly deformed.

The specimen used to produce this pattern was bent so that its surface became curved, care being taken not to fragment it and thereby introduce grain fragmentation, which would give a Laue pattern having diffuse target images. Fig. 25 is the Laue pattern

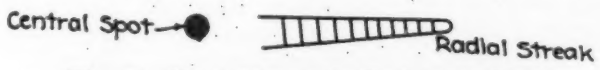


Fig. 27—Sketch Illustrating Radial Asterism.

of such a bent crystal. The "radial asterism" is now very intense, and is similar to that shown in Fig. 21. This does tend to prove that the drawn out condition of the target images must be due to the reflection of the incident X-ray beam from cylindrical or curved surfaces.

When the target images are diffuse and drawn out, then the crystals from which reflection takes place are fragmented and the atoms lie upon curved surfaces.

There is still another type of "asterism" to be considered. Fig. 15 is a spectrogram of a well-annealed steel wire. When this wire

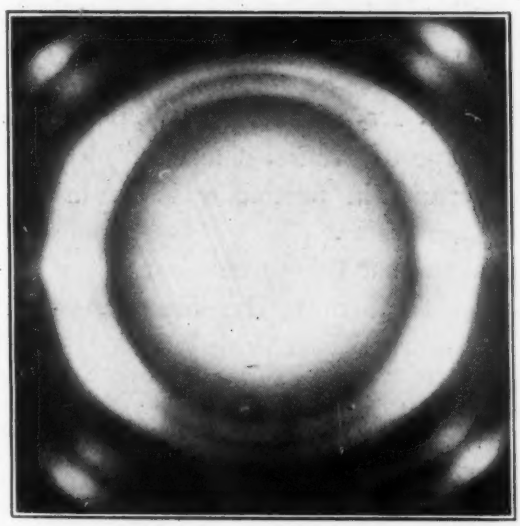


Fig. 28—Spectrogram of a Cold-rolled Molybdenum Ribbon.

was drafted 5 per cent, the pattern shown in Fig. 17 was obtained, which shows the halo surrounding the central image, intensely radial. This halo is due to the reflection of X-rays from the (1 1 0) planes. The wave length of this halo ranges from 0.30×10^{-8} to 0.50×10^{-8} centimeters. Each radial streak must be composed of a large number

of (1 1 0) planes all of which lie in the general direction of the streak, but which differ slightly in orientation from crystal fragment to crystal fragment.

Fig. 26 is an X-ray Laue pattern of what was believed to be a perfect crystal of titanium nitrocyanide. The X-ray pattern proves

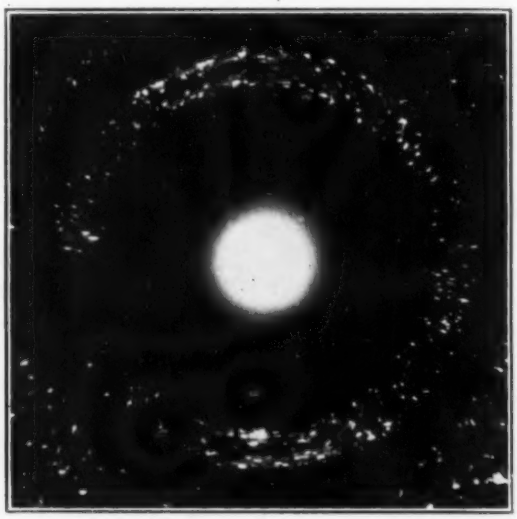


Fig. 29—Annealed Molybdenum Ribbon.

that the crystal was far from perfect. This pattern gave a clew as to the nature of this type of "radial asterism."

One of the streaks in particular is made up of a group of parallel lines as shown in Fig. 27.

This would mean that this streak is made up of crystal planes of the same kind which all lie in the same direction but are slightly oriented from one another. From this it may be possible to conclude that the "radial asterism" shown in Fig. 17 is similar to the kind shown in Fig. 26.

Effect of Annealing upon the Cold-Rolled Structure of Molybdenum Ribbon

The effects of temperature on the cold-worked structure can be studied by means of X-ray diffraction methods. Fig. 28 is a spectrogram of a molybdenum ribbon which was made by rolling an ingot down to a ribbon 0.004 inch thick. The pattern shows that the rolling caused the (1 0 0) planes to set themselves at an angle of 45 degrees with the direction of rolling, and (1 1 0) planes lie

in the direction of rolling. The (1 0 0) planes also lie in the rolling plane.

The specimen was annealed at a high temperature for a long period of time, and Fig. 29 is the result. It tells us that the structure due to cold working is not entirely removed. The rings due to the (1 1 0) planes have some orientations missing, indicating that even after a long period of annealing direction properties still exist. But the crystalline debris due to cold work is entirely removed.

Summary

To study the effect of slight plastic deformations upon the structure of steel wires, it was found that the Laue method is the most sensitive. In the early stages of grain fragmentation no orientation of the crystalline debris takes place until the grains are all broken up into fine particles of submicroscopic dimensions. After complete grain fragmentation, and with continued cold work more and more grain fragments turn so as to have (1 1 0) planes lie along the wire axis or direction of drafting.

There are two types of "radial asterism," one is due to the reflection of X-rays from curved surfaces, and the other is due to the reflection from crystal fragments which all lie in the same general direction but whose faces differ slightly in orientation from fragment to fragment.

Annealing a drastically cold-worked molybdenum ribbon causes complete recrystallization but directional properties are still evident even after weeks of heat treatment.

MECHANICAL PROPERTIES OF SOME STRUCTURAL STEELS HARDENED BELOW THE AC_3 AND ABOVE THE AR_3 POINT

By W. R. Angell

Abstract

In this paper the author has attempted to show the relationship in some structural alloy steels between normal quenching practice and low quenching; below the Ac and above the Ar points. The changes in physical properties and microstructure which take place after various heat treatments are shown. Low temperature quenching for maximum hardness and strength is of great importance when dealing with objects of uneven sections, sharp corners and small fillets. In the heat treatment of such material it is hoped that this article will be of some use to the practical heat treater.

NEW structural alloy steels are constantly appearing on the market and confront us with the problem of choosing the right heat treatment for some particular purpose. It is the object of this paper to deal with the heat treatment of some miscellaneous structural alloy steels with particular reference to low temperature quenching; below the Ac_3 and above the Ar_3 point.

It is known[1] that lag or thermal hysteresis exists in different carbon and alloy steels, although this phenomenon is much more pronounced in the latter. The elements, manganese and nickel, lower the critical points or temperatures above which maximum hardness is obtained after quenching. These critical points are considerably raised by the additions of chromium, tungsten, molybdenum, silicon and other elements. However, whether the special element which is added to the steel raises or lowers the critical points it usually increases the hysteresis or lag. By this is meant the temperature difference between the Ac_3 point on heating and the Ar_3 point on cooling, for these two points are reversals of the same thermal transformation.

[1]Birger Egeberg, "A Neglected Phenomenon in Heat Treatment," TRANSACTIONS, American Society for Steel Treating, Vol. 12, 1927, page 46.

The author, William R. Angell, member of the society, is assistant metallurgist, Naval Gun Factory, Washington, D. C. Manuscript received August 22, 1929.

Table I
Chemical Composition of Steels Investigated

Serial No.	Per Cent C	Si	S	P	Mn	Ni	Cr	V	Mo	Cu	
489	0.41	0.20	0.031	0.021	0.67	None	None	...	...	0.16	.....
283	0.45	0.10	0.027	0.027	1.09	0.18	0.06	...	...	0.09	.....
284	0.21	0.09	0.033	0.043	1.25	None	0.07	...	...	0.06	.....
451	0.51	0.23	0.024	0.029	1.35	Trace	0.08	...	...	0.02	.....
458	0.34	0.23	0.026	0.020	1.83	None	None	...	...	0.10	.....
200	0.27	0.32	0.016	0.035	0.92	3.61	...	...	...	0.27	.....
198	0.31	0.25	0.026	0.035	0.79	2.64	0.07	...	...	0.73	.....
486	0.23	0.17	0.030	0.031	0.61	4.61	...	...	....	0.32	.....
460	0.44	0.29	0.012	0.013	0.43	2.10	1.00	...	...	0.16	.....
202	0.39	0.31	0.015	0.026	0.69	2.27	1.52	...	...	0.27	.....
508	0.45	0.26	0.019	0.027	0.70	2.70	1.56	...	...	0.28	.....
266	0.42	0.20	0.031	0.008	0.25	3.03	1.32	...	...	0.08	.....
251	0.32	0.18	0.021	0.017	0.61	3.27	...	...	0.36	...	.....
490	0.55	0.23	0.022	0.016	0.76	Trace	0.93	None	None	...	.....
226	0.22	0.30	0.016	0.031	0.50	0.10	0.88	...	1.00	None	As 0.102
500	0.37	0.20	0.011	0.026	0.65	0.06	1.07	...	0.34	0.06	.....
326	0.53	0.20	0.020	0.017	0.73	...	0.80	...	0.35	...	.....
484	0.30	0.19	0.025	0.017	0.70	None	0.95	0.15	...	0.08	.....
201	0.17	0.26	0.015	0.030	0.49	0.10	0.05	...	0.96	None	As 0.097
248	0.48	1.99	0.041	0.032	0.85	None	None	...	...	0.20	.....
324	0.52	1.44	0.023	0.024	0.66	0.12	0.16	...	...	0.06	Zr 0.30
264	0.65	0.18	0.039	0.018	0.65	0.16	0.05	...	...	0.06	.....
506	0.51	0.31	0.037	0.021	1.17	0.24	None	...	...	0.04	.....
604	0.38	0.10	0.103	0.019	1.75	None	None	...	...	0.22	.....

Maximum hardness in quenched steels can be obtained by quenching above the Ar_3 point as well as the Ac_3 point. The temperature difference between these two points in some structural and special alloy steels is considerable. This is well shown in the graphs of Figs. 25 to 28 included in this paper. In these graphs hardness results are plotted for different heat treatments. For serial numbers ending in (A) hardness readings were taken from disks quenched (O indicates oil and W indicates water) at temperatures increasing by 50-degrees Fahr. increments. For serial numbers ending in (B) the disks were heated to the highest temperature shown and held long enough to become soaked through; the furnace was then cut off and the disks were quenched at the temperatures shown on the charts.

However, before advantage can be taken of the low temperature quench, just above Ar_3, the steel in question must first be heated above the Ac_3 point as in normal heat treating practice. The rate of cooling from above the Ac_3 point also has an influence on the location of the Ar_3 point. In general the faster the rate of cooling from above the upper critical the lower the depression of the Ar_3 point.

Overheating the steel considerably above the upper critical point also causes a marked depression of the Ar_3 point in some steels. Excessive overheating will have a deleterious effect on the majority of

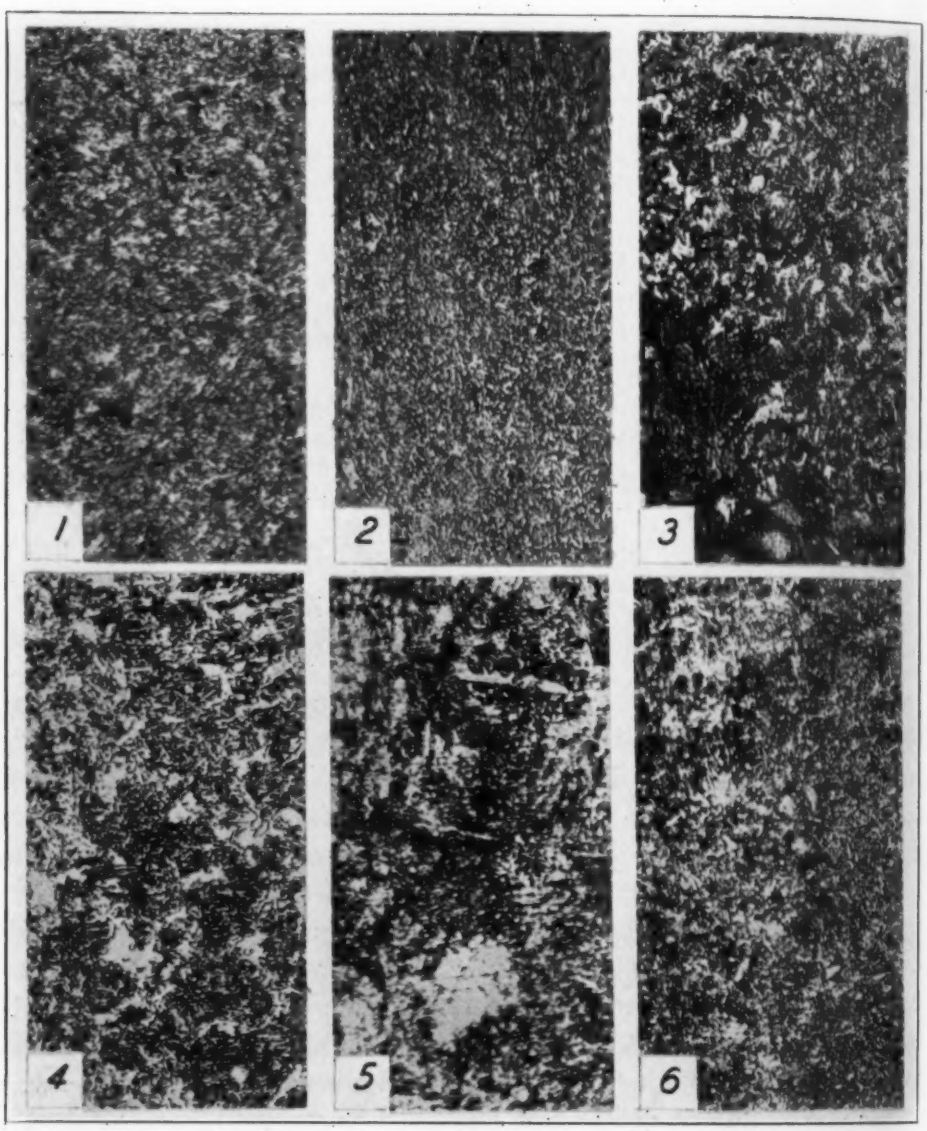


Fig. No.	Steel No.	Analysis	Heat Treatment °F.	Time	°F.	Time	
1	506	C .51; Mn 1.17	1500	30	Water	900	60 Oil
2	506	C .51; Mn 1.17	1500	30	fce cooled to 1350 Oil Qn.	900	60 Oil
3	489	C .41	1500	30	Water	900	60 Oil
4	489	C .41	1500	30	fce cooled to 1400 Water Q.	900	60 Oil
5	264	C .65; Ni .16	1500	30	Water	900	60 Oil
6	264	C .65; Ni .16	1500	30	fce cooled to 1400 Water Q.	900	60 Oil

Time in minutes. Magnification × 500.

structural steels although it may be practiced without harmful effects on such steels which react sluggishly to grain growth.

This low temperature phenomenon has been rather neglected in heat treatments of steels and should merit more study and application in our every-day heat treatments.

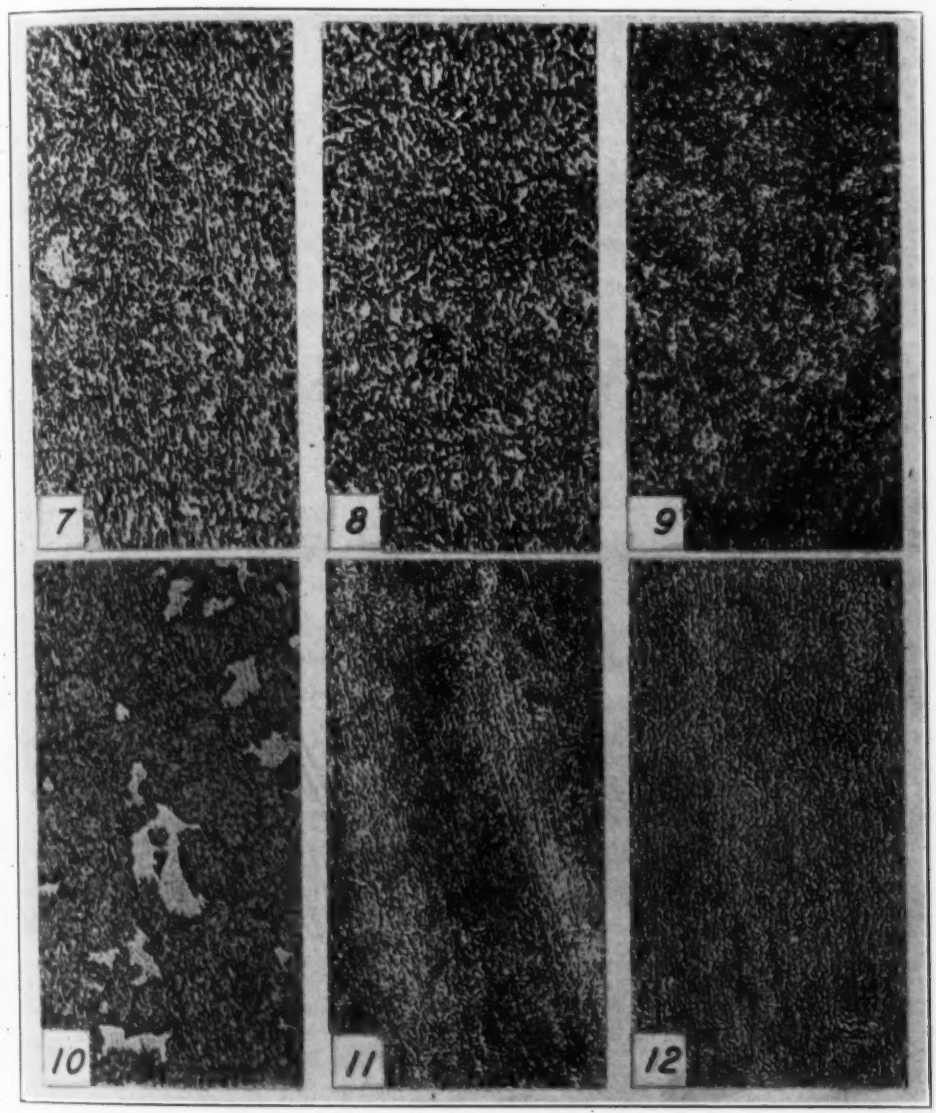


Fig. No.	Steel No.	Analysis	Heat Treatment °F.	Time		°F.	Time		°F.	Time	
7	486	C .23; Ni 4.61	1450	30	W	900	60	Oil			
8	486	C .23; Ni 4.61	1450	30	fce cooled to	1150		W. Q.	900	60	Oil
9	484	C .30; Cr .95; V .15	1600	30	W	900	60	Oil			
10	484	C .30; Cr .95; V .15	1600	30	fce cooled to	1400		W. Q.	900	60	Oil
11	326	C .53; Cr .80; Mo .35	1500	30	W	900	60	Oil			
12	326	C .53; Cr .80; Mo .35	1500	30	fce cooled to	1300		W. Q.	900	60	Oil

Time in minutes: W.—Water; O.—Oil; Q.—Quench. Magnification × 500.

Material

The material used in this investigation was supplied in 1-inch round bars, gathered from various sources. The names of the different manufacturers will be purposely omitted and reference to differ-

ent steels will be by serial and and bar number only. The chemical analysis of the steels studied are listed in Table I.

Procedure

To obtain the best quenching temperatures, 1-inch round disks one-half inch wide were quenched at temperatures beginning at 1350 degrees Fahr. and increasing in 50-degree Fahr. increments. All disks were held at quenching temperatures sufficiently long to become heated through. Where the quenching medium is water it will be understood that it was not agitated and was at room temperature. The oil tank was cooled by circulating cold water through pipes and the oil kept in motion by compressed air.

Hardness values were obtained by the Rockwell hardness tester using the 120-degree diamond cone, "C" scale and 150-kilogram load, Model "F" machine. All disks were ground down 0.020 inch on one side subsequent to quenching and the hardness taken on the ground side.

After the correct quenching temperatures were found other disks were heated to these new values and held for thirty minutes to assure complete saturation, then quenched. The current of the electric furnace was then cut off and the temperature allowed to drop (approximately 2 degrees Fahr. per minute) and at intervals of 50 degrees Fahr., except the first interval which was 100 degrees Fahr., a series of nine disks representing nine different steels was quenched. The hardness results are shown in graphical form in charts attached.

To confirm the physical results as found by the above method a set of four tension specimens and four Izod impact bars from each lot of 12 steels were heat treated. Two tension and two impact bars were quenched at the normal temperatures and another set of two tensiles and two Izods were quenched at the minimum temperature just above the Ar_3 point. The results are very satisfactory and consistent in all but one instance; serial 604 tensile tests and Izod tests 944-A and 944-B. These specimens were quenched either in the Ar_3 range or so close to the point that the desired results were not obtained, which accounts for the sudden drop in physical properties. Table II lists the physical properties of the tension and impact results; all specimens were tempered at 900 degrees Fahr. for 60 minutes prior to machining. For comparison purposes the average results of the two bars are given in the table. The tensile test specimens were standard 2-inch gage length ¾-inch thread, and the Izod

Table II

Comparison of Physical Properties of Specimens Quenched at Normal Temperatures with those Quenched Slightly Above the Ar_3 Point. All Tensile and Izod Tests were Tempered at 900 Degrees Fahr. for 60 Minutes, Followed by an Oil Quench.

Serial No.	Heat Treatment Degrees Fahr.				Prop. Limit lbs./sq. in.	Tens. Str. lbs./sq. in.	Elong. %-2 in.	Red. Area in. %	Izod Values A	Izod Values B	Hardness Values, Avg. of 3 Vickers	Hardness Values, Avg. of 3 Brinell	Hardness Values, Avg. of 3 Rock.	Hardness Values, Avg. of 3 Scler
489	1500	(30 min.)	Water		76,400	112,450	23.1	63.4	78	74	251	195	19.1	37.6
489	1500	(30 min.)	F'ce cool to 1400	(30 min.) Water	71,450	107,900	24.4	62.0	78	77	234	181	15.8	34.8
264	1500	(30 min.)	Water		105,700	158,900	14.6	39.0	6	4	372	302	35.4	46.8
264	1500	(30 min.)	F'ce cool to 1400	(30 min.) Water	101,900	156,900	15.3	40.6	8	6	384	300	35.4	46.2
604	1500	(30 min.)	Water		112,000	144,100	15.4	51.0	52	51	359	282	33.6	45.2
604	1500	(30 min.)	F'ce cool to 1250	(30 min.) Water	86,600	134,600	13.3	37.3	42	42	323	268	30.4	43.6
506	1500	(30 min.)	Water		96,950	135,350	17.6	54.9	63	62	341	269	30.9	43.6
506	1500	(30 min.)	F'ce cool to 1350	(30 min.) Water	109,500	142,750	17.1	52.6	64	62	363	285	34.4	43.6
200	1500	(30 min.)	Water		130,350	149,650	17.9	55.9	38	38	363	301	35.4	44.6
200	1500	(30 min.)	F'ce cool to 1150	(30 min.) Water	130,100	148,900	17.3	56.1	36	36	369	301	35.3	44.6
486	1450	(30 min.)	Water		127,550	142,100	18.3	58.9	48	48	339	280	33.1	44.6
486	1450	(30 min.)	F'ce cool to 1150	(30 min.) Water	123,500	141,450	18.8	58.4	47	45	334	269	32.8	41.6
202	1450	(30 min.)	Oil		165,500	191,100	12.1	40.4	19	18	434	365	43.2	54.6
202	1450	(30 min.)	F'ce cool to 1250	(30 min.) Oil	155,400	189,200	13.8	46.5	21	20	441	363	43.1	52.4
508	1450	(30 min.)	Oil		158,200	193,900	13.1	45.9	19	19	472	383	44.7	54.6
508	1450	(30 min.)	F'ce cool to 1250	(30 min.) Oil	152,800	191,900	12.8	42.7	19	21	466	383	44.4	54.0
266	1450	(30 min.)	Oil		142,600	180,600	14.5	50.7	33	32	434	352	41.6	43.2
266	1450	(30 min.)	F'ce cool to 1250	(30 min.) Oil	145,200	182,700	14.2	50.5	35	34	429	352	41.5	53.4
490	1550	(30 min.)	Water		147,700	182,500	14.5	52.4	37	40	431	359	42.2	52.0
490	1550	(30 min.)	F'ce cool to 1350	(30 min.) Water	148,000	185,400	10.7	33.0	35	37	437	363	42.8	52.6
484	1600	(30 min.)	Water		129,900	165,000	14.4	53.9	60	61	409	321	37.9	49.0
484	1600	(30 min.)	F'ce cool to 1400	(30 min.) Water	132,400	161,300	14.4	53.1	54	57	391	319	36.8	49.6
326	1500	(30 min.)	Water		163,000	213,900	2.5	13.5	18	19	499	415	46.8	57.0
326	1500	(30 min.)	F'ce cool to 1300	(30 min.) Water	163,000	213,900	8.0	19.7	21	19	509	415	46.8	60.0

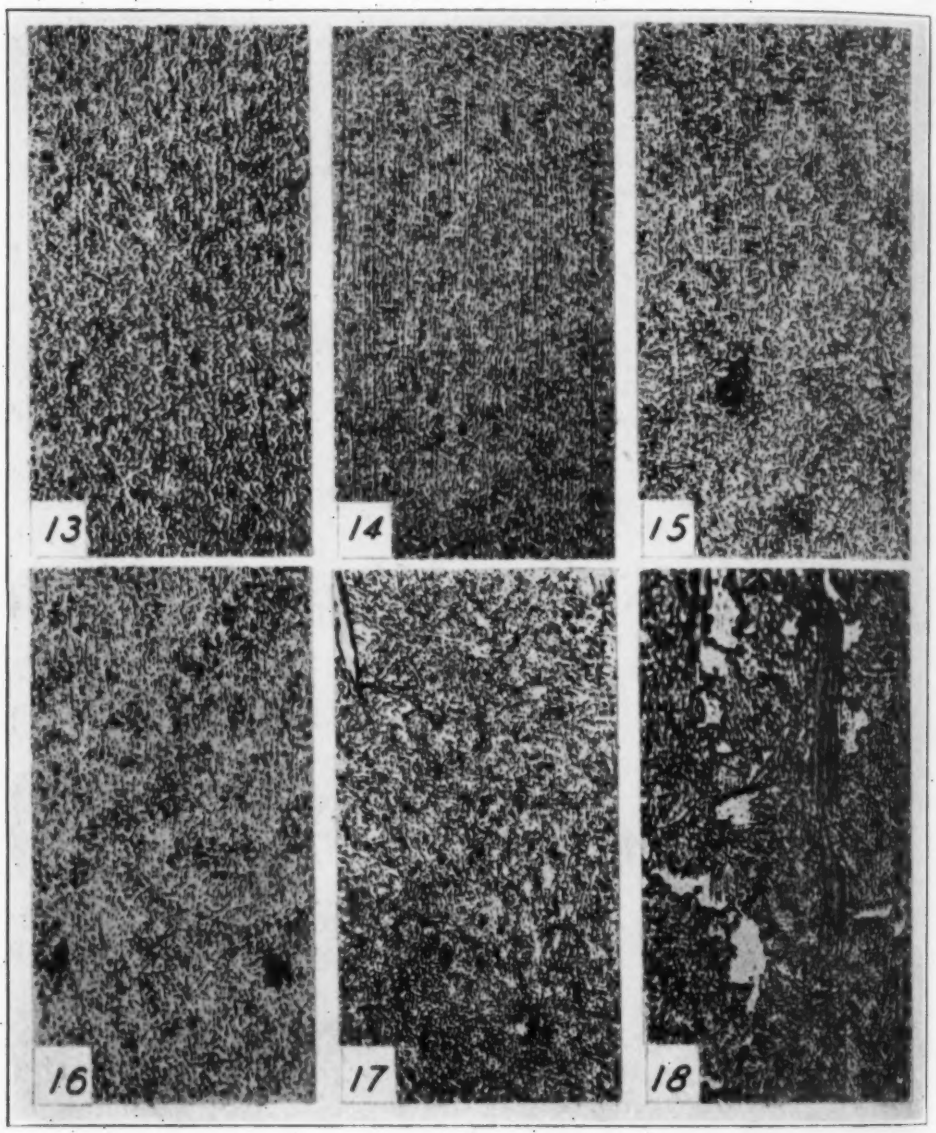


Fig. No.	Steel No.	Analysis	°F.	Time	Heat Treatment	°F.			°F.	Time	
13	202	C .39; Ni 2.27; Cr 1.52	1450	30	Oil				900	60	Oil
14	202	C .39; Ni 2.27; Cr 1.52	1450	30	fce cooled to 1250	O. Q.			900	60	Oil
15	508	C .45; Ni 2.70; Cr 1.56	1450	30	Oil				900	60	Oil
16	508	C .45; Ni 2.70; Cr 1.56	1450	30	fce cooled to 1250	O. Q.			900	60	Oil
17	604	C .38; S .103; Mn 1.75	1500	30	Water				900	60	Oil
18	604	C .38; S .103; Mn 1.75	1500	30	fce cooled to 1250	W. Q.			900	60	Oil

Time in minutes: O. Q.—Oil Quenched; W. Q.—Water Quenched. Magnification × 500.

impact bars were 10 millimeters in cross section and 4 inches long with Mesnager notch, drilled hole 2 millimeters in diameter and the center of the hole 1 millimeter below the surface, two notches to each bar.

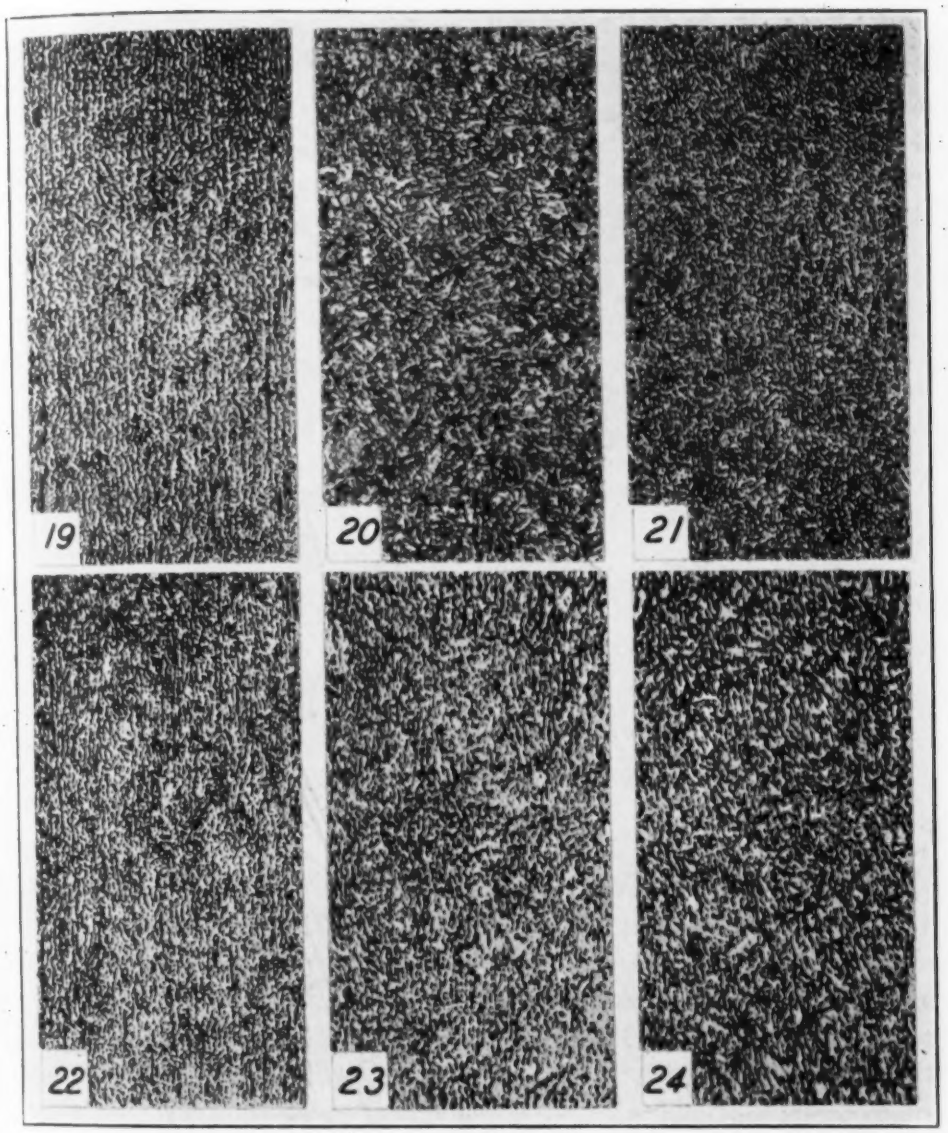
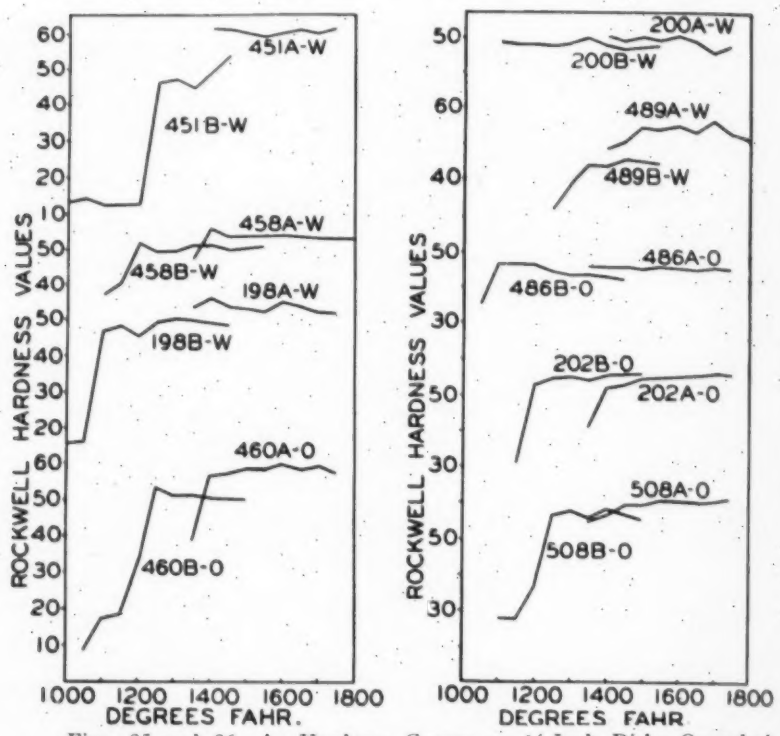


Fig. No.	Steel No.	Analysis	°F.	Time	Heat Treatment °F.		°F.	Time	
19	266	C .42; Ni 3.02; Cr 1.32	1450	30	Oil		900	60	Oil
20	266	C .42; Ni 3.02; Cr 1.32	1450	30	fce cooled to 1250	O. Q.	900	60	Oil
21	490	C .55; Cr 93	1550	30	Water		900	60	Oil
22	490	C .55; Cr 93	1550	30	fce cooled to 1350	W. Q.	900	60	Oil
23	200	C .27; Ni 3.61	1500	30	Water		900	60	Oil
24	200	C .27; Ni 3.61	1550	30	fce cooled to 1150	W. Q.	900	60	Oil

Time in minutes: O.—Oil; W.—Water; Q.—Quench. Magnification × 500.

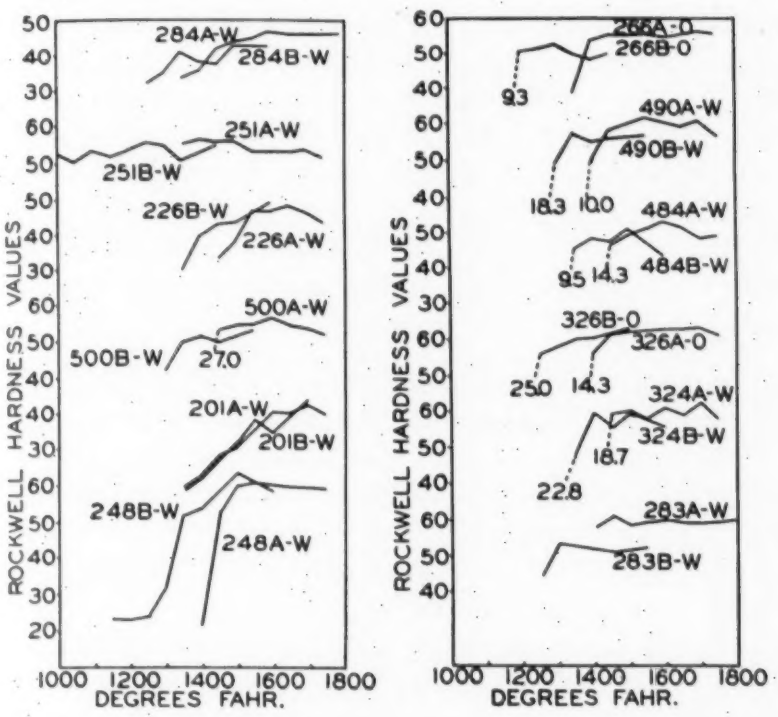
Discussion of Results

It is plain that some of the steels investigated in this report are better suited for low temperature quenching to obtain maximum physical properties than others. A glance at the various graphs included



Figs. 25 and 26—A—Hardness Curves on ½-Inch Disks Quenched in Oil or Water at Temperatures Indicated. B—Quenched at Temperatures Indicated After Being Held at Highest Temperature Shown. W—Water. O—Oil.

will make this clear. Straight carbon steels are not well suited for this type of heat treating. Steel S-489, a 0.41 per cent carbon steel, was given a 100-degree Fahr. drop from 1500 degrees Fahr. to 1400 degrees Fahr. and still retained its physical properties as compared to the normal heat treatment. Figs. 3 and 4 show this steel after the two heat treatments. Fig. 4 shows considerably more ferrite precipitated out after the 1400-degree Fahr. quench. Steel S-506 (0.15 per cent carbon and 1.17 per cent manganese) responded well to the low temperature heat treatment (tensile strength 135,000 pounds per square inch after the 1500-degree Fahr. quench as compared to 142,000 pounds per square inch after a drop of 150 degrees Fahr.). Figs. 1 and 2 show the structure after the two heat treatments, the 1350-degree Fahr. quench. Fig. 2 indicates a better structure, a finer grained sorbitic condition. Steel S-200 a 3.5 per cent nickel steel was allowed to drop from 1500 degrees Fahr. to 1150 degrees Fahr. without a loss of strength. The structure after these two treatments is shown in Figs. 23 and 24 respectively. Steel S-202 (0.39 per cent carbon—2.27 per cent nickel—1.52 per cent chromium) was dropped



Figs. 27 and 28—A—Hardness Curves on ½-Inch Disks Quenched in Oil or Water at Temperatures Indicated. B—Quenched at Temperatures Indicated After Being Held at Highest Temperature Shown. W—Water. O—Oil.

from 1450 degrees Fahr. to 1250 degrees Fahr. without showing a drop in properties. Figs. 13 and 14 respectively of this steel show about the same fine-grained structure. S-484 a 0.50 per cent carbon chromium-vanadium steel was given a drop from 1600 to 1400 degrees Fahr. without a loss of physical properties. Fig. 10 shows the microstructure after the 1400-degree Fahr. quench. Some ferrite has precipitated out of solution. This latter quenching temperature was probably dangerously close to the lower safe limit. Steel S-326 (chromium-molybdenum) gave excellent physical properties after a 200-degree Fahr. drop in quenching temperature (1500 to 1300 degrees Fahr.). Figs. 11 and 12 show the structure after each quench.

Conclusions

Low temperature quenching for maximum hardness and strength is of paramount importance when dealing with objects of intricate design. Material designed with uneven sections, sharp corners with small fillets and irregular shapes all exhibit a strong tendency to warp

and crack when quenched. The internal strains set up in such material are sometimes greater than the tensile strength of the material and the result is a crack or check, spoiling the work. A suitable tempering treatment will relieve such strains but extra precautions will have to be used in the handling of the material. Sensitive alloy steels which have a strong tendency to crack in quenching can oftentimes be successfully heat treated by taking advantage of the lower quenching temperature. Some of the conclusions which can be drawn from this investigation are as follows:

1. In all of the steels investigated it was possible to obtain as good physical properties from the lower quench as compared to the normal heat treatment.
2. The quenching range varies with the carbon content and the special elements present.
3. The physical properties of the steels used when quenched at the lower temperatures are not impaired.
4. At the lower quenching temperature the material is in a more plastic condition and can be handled more readily without deforming.
5. Low temperature quenching for hardening minimizes quenching strains.
6. Such elements as nickel, chromium, vanadium, molybdenum and manganese either by themselves or in combination with the others mentioned have a strong influence on the depression of the Ar_3 point.

Acknowledgement

The author wishes to acknowledge the courtesy of the Secretary of the Navy in permitting this data to be published and the kindly cooperation of Lieutenant Commander G. C. Hoover, and C. E. Margerum of the Naval Gun Factory. To Samuel Epstein of the Illinois Steel Company, for suggestions and helpful criticisms, and to Messrs. Lynch and Stivers of the Naval Gun Factory, for help in preparing the photomicrographs.

Recommended Practice Committee Releases*

CONSTITUTION OF IRON-PHOSPHORUS ALLOYS†

By Louis Jordan‡

THE equilibrium diagram for iron-phosphorus alloys has been investigated up to only 30 per cent phosphorus. Fig. 1 is the diagram as given by Haughton.[3] In its general form, this diagram agrees rather well with that portion of the diagram above 800 degrees Cent. and up to 21 per cent phosphorus, given by Konstantinow.[2] Konstantinow, however, indicated no field of gamma solid solution. Stead's[1] still earlier classification of iron-phosphorus alloys is consistent with Haughton's diagram. Saklatwalla's[2] constitutional diagram appears to be in error in several respects.

The iron-phosphorus liquidus falls from the melting point of iron to a eutectic of the alpha solid solution with Fe_3P at 1050 degrees Cent. and 10.2 per cent phosphorus. It then rises to a maximum of 1370 degrees Cent. at 21.7 per cent phosphorus, which is the composition of Fe_2P. During this rise the liquidus curve breaks at the peritectic temperature, 1166 degrees Cent., and at a phosphorus content slightly below 15.6 per cent, which is the compound Fe_3P. From the maximum of 1370 degrees Cent. the liquidus again falls to a second eutectic at approximately 27 per cent phosphorus and 1262 degrees Cent., which is a eutectic between Fe_2P and a constituent as yet unidentified.

Haughton determined the limits of solid solubility of phosphorus in iron between the melting point of iron and the eutectic line at 1050 degrees Cent. by heat treatment and microscopic examination of the specimens, and also from the eutectic line down to 600 degrees Cent. by both metallographic examination and thermal analysis. The solubility of phosphorus at the eutectic temperature is 2.8 per cent and at the A_2 point is 1.1 per cent (A_2 is at 745 degrees Cent. in a 1.1 per cent phosphorus alloy).

The solution of phosphorus in iron rapidly lowers the A_1 transformation and raises the A_3. The two lines join at about 1050 degrees Cent. and 0.5 per cent phosphorus which gives an enclosed field of gamma solid solution.

Magnetically determined heating and cooling curves show the A_2 transformation in all alloys up to 13 per cent phosphorus and a second

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‡Senior Chemist, Division of Metallurgy, National Bureau of Standards, Washington, D. C.

The releases from the Recommended Practice Committee as printed herewith will be included in the next edition of the A. S. S. T. HANDBOOK, which will be published about June 1, 1930.

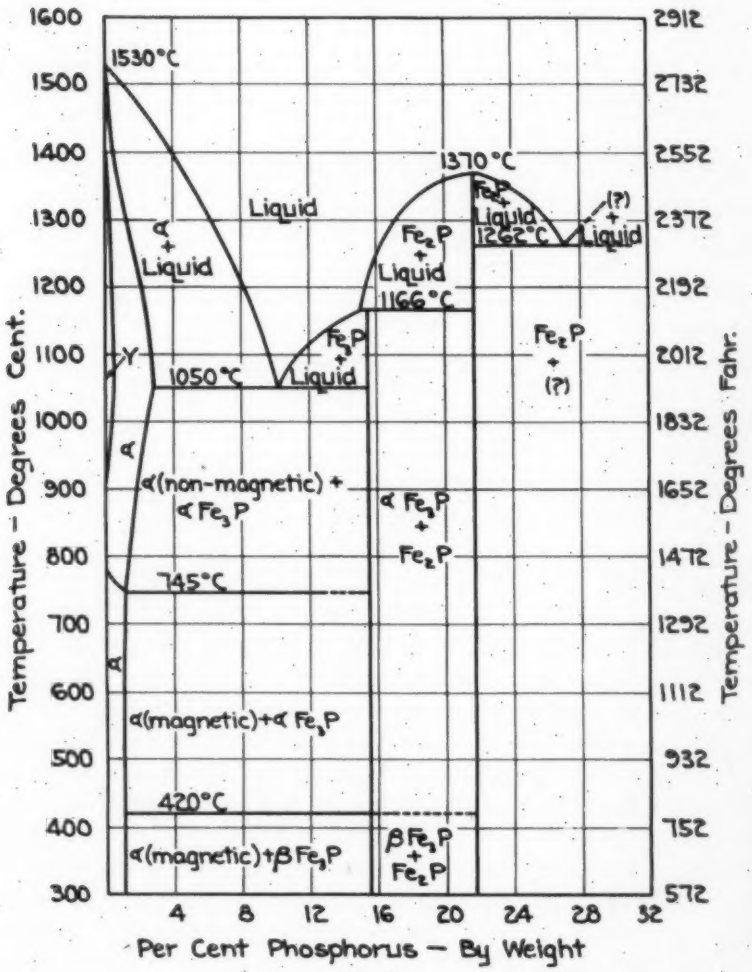

Fig. 1—Iron-Phosphorus Equilibrium Diagram (Haughton)

magnetic point at 420 degrees Cent., which is attributed to a transformation in Fe_3P similar to the A_0 in Fe_3C. Consequently this magnetic point at 420 degrees Cent. is indicated in the diagram as continuing to the composition Fe_2P.

Hägg[9, 10] from X-ray analyses has assigned to the compound Fe_3P a body-centered tetragonal lattice and to Fe_2P a hexagonal lattice of the dimensions a = 5.852 Å and c = 3.453 Å. He also found some evidence indicating the existence of FeP. Friauf[11] has recently confirmed Hägg's results for the structure and lattice dimensions of Fe_2P.

Ferrophosphorus—Ferrophosphorus has some use as an addition to certain open hearth steels which are to be rolled to sheets, particularly for light gages. A small amount of phosphorus in the steel tends to prevent sticking of the sheets. Ferrophosphorus also may be added to a cupola charge when the pig iron is too low in phosphorus to meet the casting requirements. Two grades of the alloy are made, ([12, 13, 14]) a 17 to

19 per cent alloy and a 23 to 25 per cent alloy. The latter is an electric furnace product while the former may be produced in the blast furnace.

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TEMPERATURE MEASUREMENTS IN MOLTEN ALUMINUM AND ITS ALLOYS

By Kirtland Marsh†

Type of Pyrometer—A thermoelectric pyrometer provides the most satisfactory means for determining the temperature of molten aluminum. The subject of thermoelectric pyrometers has been discussed in other sections of the handbook, so only those phases of the subject pertaining particularly to the measurement of the temperature of molten aluminum will be discussed.

Cold Junctions—Since close temperature control is desirable (± 20 degrees Fahr. in the melting operation and ± 10 degrees Fahr. in the pouring operation), provision should be made for maintaining a constant cold junction temperature for all thermocouples, or means must be provided to compensate for fluctuating cold junction temperatures. This is easily accomplished and is mentioned only because it is a requirement which is too often neglected altogether, or only partially met. Fluctuation of the cold junction temperature of a base metal thermocouple will produce errors in the temperature reading of about the same magnitude as the fluctuation of the cold junction temperature.

Thermocouple Protection Tubes—Every base metal thermocouple in common use is readily attacked by molten aluminum, some to a greater extent than others, but all to a sufficient extent to warrant the use of a suitable protection tube over the thermocouple wherever practicable. A thermocouple actually measures the temperature of its hot junction, so that the protection tube and the hot junction must become the same temperature as the material before a correct reading can be obtained. A protection tube over a thermo-

†U. S. Aluminum Co., New Kensington, Pa.

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couple materially increases the time required to secure a correct temperature reading. In the case of a melting furnace, the thermocouple and protection tube can be immersed as soon as the aluminum is melted down and allowed to remain in the metal at least until it is ready to be poured, so that the five minutes or more required for the protection tube and thermocouple to reach the temperature of the molten metal is not a handicap. However, when it is desired to measure the temperature of the metal in a pouring ladle, time is an important factor and a protection tube is out of the question.

Protection tubes may be made of iron or steel pipe, gray cast iron, or some of the heat resisting alloys available on the market. Even the heat resisting alloys are not effective in resisting the alloying or dissolving action of molten aluminum, so it is desirable to apply a protective coating or wash to the tube. A protection tube made of gray cast iron will last for several months in a melting pot if properly protected, and since the failure of a tube is so frequently due to the accidental rupture of the protective coating the cheaper cast iron tubes are generally the most economical.

Protective Coatings for Tubes—A simple and inexpensive method of preventing the molten aluminum from dissolving or alloying with the protection tube consists of applying a coat of whitewash made up of hydrated lime and water mixed in equal proportions by volume. The wash can be easily applied by a brush, particularly if the tube is heated to a temperature under the boiling point of water (180 to 200 degrees Fahr.). Another method is to dip the heated protection tube once into the wash and quickly withdraw it before the tube is sufficiently chilled to prevent the rapid drying of the wash.

If the thermocouple is for use in an open-top melting pot and is suspended from a point above the pot, the above wash is very effective, but if the thermocouple is mounted in such a manner that the protection tube is subject to contact with solid objects, the wash will be easily abraded or knocked off. In such cases, a protective coating much more resistant to abrasion can be made by mixing the lime with a solution consisting of 25 per cent by volume of commercial sodium silicate and 75 per cent water. This wash can be applied in the same manner as the lime and water wash, but the sodium silicate content should not be allowed to increase more than a few percent by evaporation of the water, or a spongy and less satisfactory coating will result.

Protection tubes should be recoated as often as necessary to maintain an effective coating. Before a tube is recoated, it should be carefully cleaned to remove any aluminum which may cling to it, and also any previous coating which may have become loosened. Occasionally the entire tube should be scraped clean to remove any previous coat and any loose iron oxide scale, since a thick coat of lime or loose scale will readily crack off while the tube is being heated up in the molten metal. The lime wash is much more adherent to a new cast iron protection tube which has been exposed to atmospheric influence long enough to become rusted, than to a smooth unrusted casting. The rusting should not be allowed to progress far enough to produce a loose scale.

Thermocouple Holder—A large proportion of aluminum is melted in open-top melting pots. It is generally a simple matter to suspend a thermocouple over the pot so that it may be lowered into the metal or raised out of

the way as required. A satisfactory protection tube and holder for open-pot melting furnaces are shown in Fig. 1. The holder may be suspended by means of a chain attached by screws at each end. The tube is readily removed from the holder so that replacement is simple.

Thermocouples—The measurement of the temperature of molten aluminum in pouring ladles must be accomplished easily and quickly so that

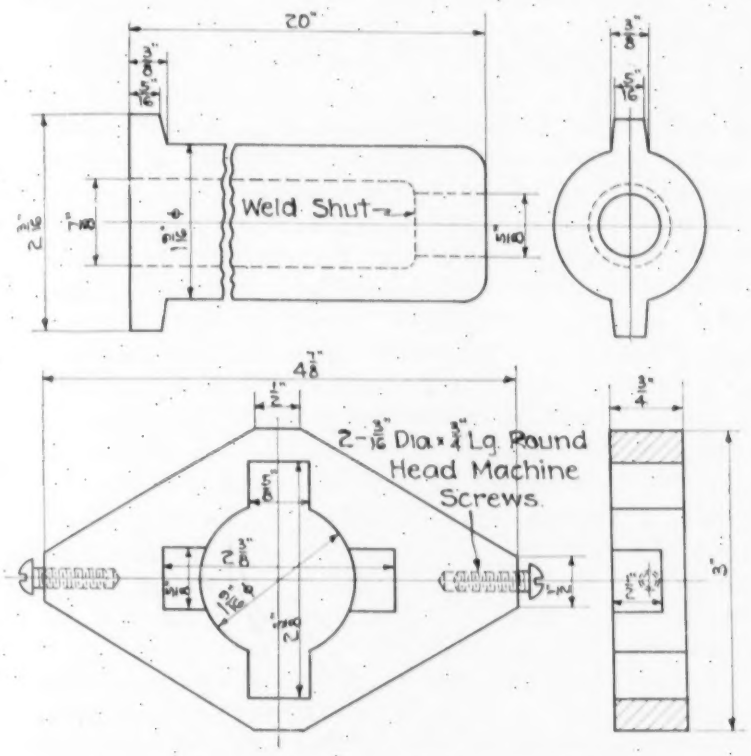

Fig. 1—Protection Tube and Holder for Thermocouple.

the use of a thermocouple in a heavy protection tube is out of the question. Furthermore, many thermocouples themselves are sufficiently massive to require an appreciable length of time to attain temperature equilibrium in a bath of molten aluminum.

The most common form of thermocouple consists of two wires of different chemical composition welded together at one end to form the hot junction. The temperature actually measured is the temperature of the hot junction. The correct temperature of the molten aluminum is, therefore, not obtained until the thermocouple has been in the molten metal long enough for the hot junction to reach the temperature of the molten metal. If, however, the two wires are not welded, the molten aluminum will complete the electrical circuit and the true temperature of the molten metal will be indicated in a much shorter time than in the case of a welded thermocouple made of wires of the same size or mass.

A quick reading thermocouple can be made of No. 8 B. & S. gage nickel-

chromium thermocouple wire. Both wires should be insulated with asbestos to within ¼ to ½ inch to the end, otherwise the hot junction between the two wires may be formed at the surface of the molten aluminum. With the elements insulated, the hot junction will be at or close to the ends of the elements, so the temperature at any depth in the pouring ladle can be obtained by immersing the couple to that depth.

Thermocouple Insulations—Not all types of asbestos insulation are satisfactory. The best form is a continuous felted layer of asbestos on each wire. This can be obtained from the pyrometer manufacturers.

The felted asbestos will become brittle and rub off, but with reasonable care the wear on the asbestos insulation will about equal the erosion of the wires. If the insulation wears away faster than the wires are dissolved, the bare wire should be cut off to within ¼ inch of the insulation, so that the hot junction, formed by the molten metal, is well below the surface.

Cleaning Thermocouples—Most of the molten aluminum which may adhere to the thermocouple when it is withdrawn should be shaken off before the aluminum solidifies; otherwise the next time the couple is used a much longer time will be required to obtain a true temperature reading.

The thermocouple should never be whipped against any solid object to remove the molten aluminum, because that would break off the insulation. To remove the aluminum, move the couple rapidly in a direction parallel to its length and then suddenly stop it; similar to shaking ink from a fountain pen.

Checking Pyrometers—The best pyrometers must be checked occasionally and they should be checked or calibrated at about the same temperature for which they are used. Because of the definite freezing or arrest points of pure aluminum and its alloys, it is a simple matter to calibrate the pyrometers in an aluminum foundry, as follows:

Fill a small crucible or pouring ladle with molten pure aluminum or one of the alloys shown in Table I, and immerse the thermocouple to a depth of 6 to 8 inches in the molten metal. While gently stirring the molten metal with the thermocouple or a stirring rod, read the temperature as indicated by the instrument at ½ minute intervals. The temperature will drop at a relatively uniform rate until the freezing point of the metal is reached, when the drop in temperature will cease or become much reduced until the metal is entirely solidified. The temperatures given in Table I apply only to the beginning of the arrest period or to the point at which the direction of the time-temperature curve changes. If the mass of molten metal were great enough and it were vigorously stirred there would be a period during which the temperature would not change, but in practice there is generally only a change in the slope of the time-temperature curve. Enough metal should be used, stirring should be sufficiently vigorous, and the depth of immersion of the thermocouple in the molten aluminum great enough that the change in slope of the curve will be sharp and well defined, rather than gradual or rounded off. No. 12 alloy does not give as sharp a change in slope as pure aluminum or the other alloys mentioned, but can be used if no other metal is available. The temperature as indicated by the pyrometer should agree with the temperature as given in the table, at the beginning of the arrest period,

but if it does not the pyrometer is in error by an amount equal to the difference between the true temperature and the temperature as indicated.

Although a pyrometer in an aluminum foundry is not ordinarily used at temperatures as low as the freezing point of aluminum, the temperature at which it is used is sufficiently close to the freezing temperature, so that the error will be about the same.

In order to obtain a definite arrest point, the metal in the crucible should be stirred continuously. A heavy protection tube cannot be used over the couple, for this would almost entirely obliterate the arrest point on the time-temperature curve. The quick reading couple can be immersed directly in the molten aluminum, and the welded couples used in protection tubes in the melting pots can be immersed in the same manner if they are first given a protective coating of lime wash. The coating should be perfectly dry before the couple is immersed, and the coating must not be rubbed off while the couple is in the metal.

Not all alloys of aluminum have a sufficiently definite arrest period during solidification to make them suitable for freezing point calibrations. Furthermore, since the arrest period occurs at different temperatures for different alloys, the metal selected must be of known composition or the

Table I
Freezing Points of Aluminum and Its Alloys

Composition by Per Cent									
Al	Cu	Si	Zn, max.	Fe, max.	Mg	Mn, max.	Total Impurities	Alloy	Arrest Point Degrees Fahr.
99.95	.....	.....	...	...	..	...	.05	Pure Al	1220
99.2	.....	.....	...	...	..	...	.8	Comm. Pure Al	1216
67.0	33.0	.....	...	...	..	...	.05	Al-Cu Eutectic[1]	1018
67.0	33.0	.....	...	...	..	...	.8	Al-Cu Eutectic[2]	1010
91.1-89.6	7.0-8.5	.....	0.2	...	..	....	1.7	No. 12	1160
86.15-85.15	0.15†	12.0-13.0	0.2	0.8	*	0.7	...	No. 47 Not Modified	1071

[1]Made from pure (99.95 per cent) aluminum.
[2]Made from commercially pure (99.2 per cent) aluminum.
†Maximum. *Trace.

freezing temperature must be determined experimentally with a pyrometer of known accuracy. It is generally possible to secure enough new metal, either commercially pure aluminum or one of the alloys mentioned in the table, for a freezing point determination.

It is not necessary to allow the metal to completely solidify in making a freezing point determination, for with a reasonable amount of metal, the arrest period will extend over several minutes, and as soon as it is definitely reached the couple may be withdrawn and the metal dumped out of the crucible before it becomes solid.

If the zero setting of the instrument is properly maintained, any change in calibration would ordinarily be gradual unless due to accident, in which case the error would probably be large enough to be apparent without an actual

calibration. In some large aluminum foundries, the quick reading couples for taking pouring temperatures are checked once a day. The writer recommends the calibration of all quick reading couples at least once a week, and of all others, at least once a month.

CONSTITUTION OF IRON-MOLYBDENUM ALLOYS

By W. P. Sykes*

The most recent and complete reports on studies of this alloy system are those published by Sykes,[1] Takei and Murakami,[2] and Arnfelt.[3] The constitutional diagram in Fig. 1 is drawn from the data included in these

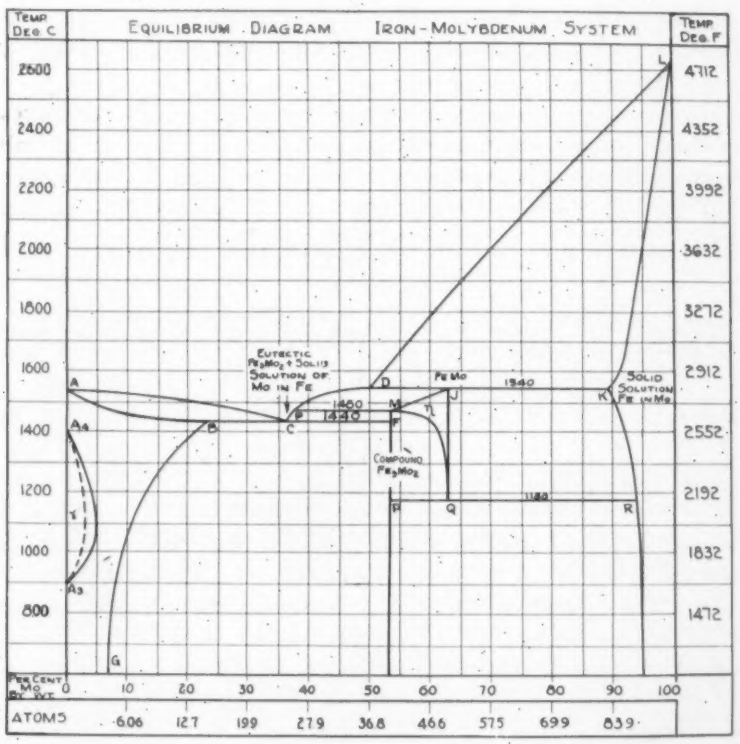

Fig. 1—Iron-Molybdenum Equilibrium Diagram.

reports. As this diagram indicates, the system may be described as follows:

(a) A solid solution of molybdenum in iron with about 24 per cent molybdenum soluble at 1440 degrees Cent. and about 6 per cent at room temperature.

(b) By the addition of about 3 per cent molybdenum to iron the Ar_4 point is lowered from 1400 to 1230 degrees Cent. and the Ar_3 point raised from 890 to 980 degrees Cent.

*General Electric Co., Incandescent Lamp Dept., Cleveland Wire Works, Cleveland, Ohio.

(c) A eutectic at about 36 per cent molybdenum, which is a conglomerate of iron-rich solid solution and a phase corresponding to the composition Fe_3Mo_2. The temperature of this eutectic lies between 1440 and 1450 degrees Cent.

(d) The phase Fe_3Mo_2 decomposes upon heating above about 1480 degrees Cent. to form an iron-rich liquid and another solid, the eta phase.

(e) The eta phase appears to have a maximum molybdenum content of about 63 per cent. This corresponds to the composition of an intermetallic compound having the formula FeMo. The region MPQJ in the

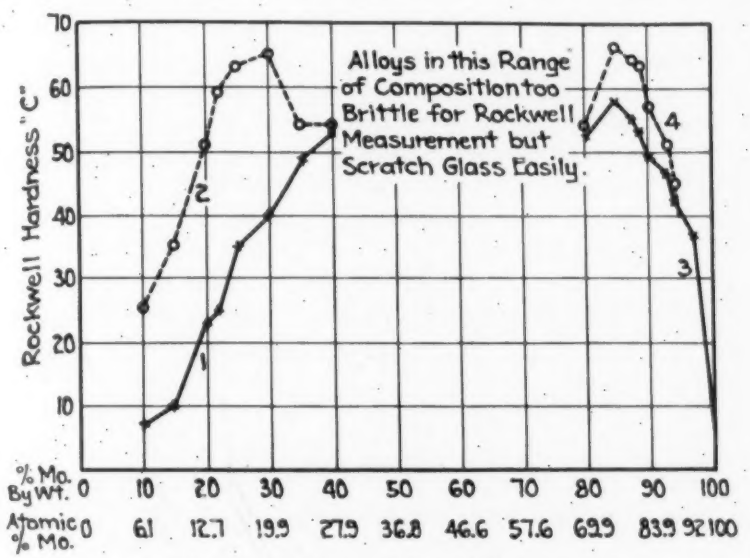

Fig. 2—Rockwell Hardness of Iron-Molybdenum Alloys.

diagram may be considered to be occupied by two phases, the Fe_3Mo_2 and the eta phase. When an alloy whose composition lies between 53 and 63 per cent molybdenum is slowly cooled through a temperature lying on the line MQ, Fe_3Mo_2 is precipitated from the eta phase.

(f) The eta phase in cooling undergoes a eutectoid transformation at about 1180 degrees Cent. (PQR) the products being Fe_3Mo_2 and a molybdenum-rich solid solution containing about 6 per cent iron. When heated above 1540 degrees Cent. the eta phase decomposes into an iron-rich liquid and a solid solution of molybdenum containing about 11 per cent iron.

(g) The solid solubility of iron in molybdenum decreases from about 11 per cent at 1540 degrees Cent. to about 5 per cent at room temperature.

Properties of Pure-Iron-Molybdenum Alloys

The Rockwell and Brinell hardness numbers of the iron-rich alloys increase continuously with the addition of molybdenum up to about 40 per cent by weight of molybdenum. Pure iron as quenched from 1425 degrees Cent. has a Brinell hardness of about 60. An alloy containing 40 per cent

of molybdenum quenched from the same temperature has a Brinell hardness of about 450.

Compositions lying between 40 and 80 per cent of molybdenum are composed largely of the intermetallic compounds and are glass-hard, but very brittle. Compositions within the solid solutions ranges at either end of the system are markedly susceptible to precipitation hardening as shown by the curves in Fig. 2. Those alloys containing about 20 per cent molybdenum will develop a Brinell hardness of 500 to 550 by aging at 625 degrees Cent. for 20 hours after having been quenched from 1400 degrees Cent. The hardness so developed will persist over a long period at temperatures up to 650 degrees Cent.

On account of this high temperature hardness alloys in the range of 20 per cent molybdenum have been found to give excellent service when used for dies in hot swaging operations.

Due to their normal coarse-grained structure, however, these alloys are relatively brittle and consequently, in their present stage of development, have a decidedly limited field of application.

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PHYSICAL CONSTANTS OF LEAD

By James E. Harris†

Atomic Weight—The atomic weight of lead is 207.22. Lead is unique among the commonly occurring elements in that specimens have been found having atomic weights differing materially from the usually accepted value. These samples have in all cases been found associated with radio-active materials and have undoubtedly been formed by the radio-active disintegration of elements of higher atomic weights. Determination of the atomic weights of samples of lead from such sources by O. Honigschmid and co-workers and by T. W. Richards and co-workers have given values ranging from 206.046[1] to that of ordinary lead. Two determinations, those of Soddy and Hyman[2] and of O. Honigschmid[3] give values higher than that of ordinary lead; namely, 208.4 and 207.92 respectively.

Atomic Number—The atomic number of lead is 82.

Isotopes—A. S. Russell[4] has derived the hypothetical conclusion that lead consists of isotopes of atomic masses of 204, 205, 206, 207, 208, and 210. F. W. Aston[5] has found mass spectrum lines corresponding to atomic weights 206, 207, and 208. He found indications of an isotope at 209 and possibly at 203, 204, and 205.

†Research Metallurgist, Bell Telephone Laboratories, New York.

Allotropic Modifications—Cohen and Helderman,[6] Heller,[7] Janecke,[8] and others have come to the conclusion that lead exists in two allotropic modifications with a transition point at about 50 degrees Cent. This conclusion is based in part on the fact that when lead is used as a cathode in a solution of lead acetate and nitric acid there are marked structural changes, the volume increasing slightly, the color changing to gray, the metal losing its malleability and firmness. Janecke by measuring the pressure-temperature cooling and heating curves found indications of a transition point between 59 and 62 degrees Cent. Bridgman[9] found no evidence of allotropism in his observations of the thermo-electric properties of lead. A. Thiel[10] concludes that the results obtained by Heller, Cohen and others are due to chemical changes since the disintegration takes place only in the presence of nitric acid. Rawdon[11] could find no evidence of the existence of lead in allotropic modifications.

Density—Density at 20 degrees Cent. of lead cast in vacuo is 11.3437 ± .0001.[12] Density of solid lead at the melting point is 11.0.[13] The density of molten lead is given as follows:[14]

Degrees, Cent.	327	350	450	550	650	750	850
Density	10.686	10.658	10.536	10.418	10.302	10.188	10.078

The density of lead from radio-active materials has been found to differ from that of ordinary lead in such a way as to make the atomic volume constant.

Crystal Form—Lead crystallizes in the face-centered cubic lattice with the edge of length 4.920 ± .005Å.[15]

Melting Point—The melting point of lead is 327.35 degrees Cent. ± .05 degrees.[16]

Boiling Point—The boiling point of lead is variously given from 1525 to 1870 degrees Cent. J. Johnston[17] computes the boiling point from vapor pressure data at 1640 and Rodebush and Dixon[18] at 1619 degrees Cent. H. C. Greenwood[19] and W. R. Mott[20] give 1525; Ruff and Bergdall,[21] 1555; H. Von Wartenburg,[22] 1580 ± 20; P. Walden,[23] 1700; and J. J. Van Laar,[24] 1870 degrees Cent. as the boiling point of lead.

H. C. Greenwood gives the boiling points at different pressures as follows:

Pressure in Atmospheres	.14	.35	1.0	6.3	11.7
Boiling Point, Degrees Cent.	1325	1410	1525	1873	2100

Critical Temperature and Pressure—J. J. Van Laar[24] estimates the critical temperature as 3000 degrees Cent. and the critical pressure as 370 atmospheres. P. Walden[23] gives 3598 and W. Herz,[25] 2696 degrees Cent. as the critical temperature.

Heat of Fusion—The heat of fusion of lead is 5.5 calories per gram.[26]

Heat of Vaporization—Rodebush and Dixon[27] gives 223 calories per gram for the heat of vaporization of lead.

Specific Heat

Specific heat[29] (mean) —180 to 20 degrees Cent.....0.0300
Specific heat[30] (mean) 0 to 100 degrees Cent.....0.03046

Naccari[31] gives the following values for the specific heats at the temperatures indicated:

Degrees, Cent.	15	50	100	150	200	250	300
Specific Heat	0.02993	0.03040	0.03108	0.03176	0.03243	0.03312	0.03380

Dixon and Rodebush[32] give the following values in calories per gram for the specific heats of liquid lead at the temperatures indicated:

Degrees, Cent.	365	378	418	459
Specific Heat	0.0335	0.0338	0.0335	0.0335

Thermal Conductivity of Lead—S. Konno[33] gives the following values for the thermal conductivity of lead. The values are given in calories per centimeter per second.

Degrees, Cent.	0	100	200	400	500	600	700
Thermal Conductivity	0.083	0.081	0.077	0.038	0.037	0.036	0.036

Thermal Expansion of Lead

Linear coefficient of expansion[34] per degrees Cent. (mean) 10 to 100 degrees Cent.....29.5 x 10^{-6}

Linear coefficient of expansion[35] per degrees Cent. (mean) −190 to 19 degrees Cent.....26.5 x 10^{-6}

Cubical coefficient of expansion[36] at 20 degrees Cent.....85 x 10^{-6}

Electrical Resistivity—Schumacher and Bouton[37] give the following data for the resistivity of lead. The values are in micro-ohms per cubic centimeter.

Degrees, Cent.	25	40	70	100	146
Electrical Resistivity	21.3	22.5	25.2	27.6	31.5

For temperatures above the melting point Matuyama[38] gives

Degrees, Cent.	348	365	420	472	527	682	731	776	856
Electrical Resistivity	96.9	97.6	100	103	108	112	114	117	120

For low temperatures, H. K. Onnes and co-workers give in values referred to the resistivity of lead at zero as unity.

Degrees, Cent.	−103.63	−195.15	−216.61	−258.70
Electrical Resistivity	0.59548	0.25257	0.17129	0.01311

At the temperature of liquid helium lead becomes a superconductor.

Temperature Coefficient of Electrical Resistivity—The temperature coefficient of electrical resistivity of lead is given by Bridgman[39] for the range 0 to 100 degrees Cent. as .00421.

Thermal E. M. F. of Different Elements Against Lead—The values for the constants in the equation $\frac{dE}{dt} = A + Bt$ are given in the Table I. The thermoelectric powers at 20 and 50 degrees Cent. and the neutral points are also given. The values given have been compiled by F. E. Fowle[40] from data presented by Becquerel, Mathiessen and Tait.

Electrode Potential—G. N. Lewis and T. B. Brighton give the normal

Table 1

The Thermoelectric Power of Lead Against Other Metals

Substance	A Micro-volts	B Micro-volts	Thermoelectric Power at Mean Temp. of Junctions (Microvolts) 20° C.	50° C.	Neutral Point
Aluminum	− 0.76	+0.0039	− 0.68	− 0.56	+ 195
Antimony, commercial pressed wire	.....	.....	+ 6.0	.....	.....
Antimony, axial	.....	.....	+ 22.6	.....	.....
Antimony, equatorial	.....	.....	+ 26.4	.....	.....
Antimony, ordinary	.....	.....	+ 17.0	.....	.....
Argentan	−11.94	−0.0506	− 12.95	− 14.47	− 236
Argentan	.....	.....	.....	− 12.7	.....
Arsenic	.....	.....	− 13.56	.....	.....
Bismuth, commercial pressed wire	.....	.....	− 97.0	.....	.....
Bismuth, pure pressed wire	.....	.....	− 89.0	.....	.....
Bismuth, crystal, axial	.....	.....	− 65.0	.....	.....
Bismuth, crystal, equatorial	.....	.....	− 45.0	.....	.....
Bismuth, commercial	.....	.....	.....	− 39.9	.....
Cadmium	+ 2.63	+0.0424	+ 3.48	+ 4.75	− 62
Cadmium fused	.....	.....	.....	+ 2.45	.....
Cobalt	.....	.....	− 22.	.....	.....
Constantan	.....	.....	.....	− 19.3	.....
Copper	+ 1.34	+0.0094	+ 1.52	+ 1.81	− 143
Copper commercial	.....	.....	+ 0.10	.....	.....
Copper galvanoplastic	.....	.....	+ 3.8	.....	.....
Gold	.....	.....	+ 1.2	.....	.....
Gold	+ 2.80	+0.0101	+ 3.0	+ 3.30	(− 277)
Iron	+17.15	−0.0482	+ 16.2	+ 14.74	+ 356
Iron pianoforte wire	.....	.....	+ 17.5	.....	.....
Iron commercial	.....	.....	.....	+ 12.10	.....
Iron commercial	.....	.....	.....	+ 9.10	.....
Lead	.....	.....	.....	.....	.....
Magnesium	+ 2.22	−0.0094	+ 2.03	+ 1.75	+ 236
Mercury	.....	.....	− 0.413	.....	.....
Mercury	.....	.....	.....	− 3.30	.....
Nickel	.....	.....	.....	− 15.50	.....
Nickel (−18 to 175° C.)	−21.8	−0.0506	− 22.8	− 24.33	(− 431)
Nickel (250 to 300° C.)	−83.57	+0.2384	.....	.....	.....
Nickel (above 340° C.)	− 3.04	+0.0506	.....	.....	.....
Palladium	− 6.18	−0.0355	− 6.9	− 7.96	− 174
Palladium	.....	.....	.....	− 6.9	.....
Phosphorus (red)	.....	.....	+ 29.9	.....	.....
Platinum	.....	.....	+ 0.9	.....	.....
Platinum (hardened)	+ 2.57	−0.0074	+ 2.42	+ 2.20	347
Platinum (malleable)	− 0.60	−0.0109	− 0.818	− 1.15	− 55
Platinum wire	.....	.....	.....	+ 0.94	.....
Platinum another specimen	.....	.....	.....	− 2.14	.....
Platinum iridium alloys:					
85% Pt+15% Ir	+ 7.90	+0.0062	+ 8.03	+ 8.21	(−1274)
90% Pt+10% Ir	+ 5.90	−0.0133	+ 5.63	+ 5.23	444
95% Pt+ 5% Ir	+ 6.15	−0.0055	+ 6.26	+ 6.42	(−1118)
Selenium	.....	.....	+807.	.....	.....
Silver	+ 2.12	+0.0147	+ 2.41	+ 2.86	− 144
Silver (pure hard)	.....	.....	+ 3.00	.....	.....
Silver wire	.....	.....	.....	+ 2.18	.....
Steel	+11.27	−0.0325	+ 10.62	+ 9.65	347
Tellurium	.....	.....	+502.	.....	.....
Tellurium	.....	.....	.....	+429.3	.....
Tellurium β	.....	.....	+500.	.....	.....
Tellurium α	.....	.....	+160.	.....	.....
Tin (commercial)	.....	.....	.....	+ 0.33	.....
Tin	.....	.....	+ 0.1	.....	.....
Tin	− 0.43	+0.0055	− 0.33	− 0.16	78
Zinc	+ 2.32	+0.0238	+ 2.79	+ 3.51	− 98
Zinc pure pressed	.....	.....	+ 3.7	.....	.....

electrode potential of lead against the normal calomel electrode as 0.4125 volt and against the normal hydrogen electrode as 0.1295 volt.[41]

Surface Tension—The following values from 327 to 500 degrees Cent.

are given by T. R. Hogness[42] and from 750 to 1000 degrees Cent. by Drath and Sauerwald.[43] The values given are in dynes per centimeter.

Degrees, Cent.	327	400	500	750	900	1000
Surface Tension ..	444	438	431	423	401	397

Capillary Constant—Drath and Sauerwald[43] give the value 8.37 for the capillary constant.

Viscosity—The following values for the coefficient of viscosity of molten lead are given by R. Arpi.[44] The values are in C. G. S. units.

Degrees, Cent.	340	376	419	470
Viscosity	0.0189	0.0167	0.0160	0.0144

Optical Properties—H. von Wartenburg[45] gives for the index of refraction of lead for wave length 0.589μ the value 2.01; for the absorption index 3.48; and for the reflecting power 62 per cent.

A. K. Aster[46] gives for the index of refraction of molten lead for a wave length of approximately 0.602μ the value 0.415 and for the absorption coefficient 1.76. Both were found to be constant over the temperature range 335 to 485 degrees Cent.

Tensile Strength, Elongation, and Hardness—The following data are given by Hiers[47] for pure lead and two grades of commercial lead in the cast form.

	Southeastern Missouri Desilverized	Chemical Lead	Very Pure Lead
Yield point (0.5 per cent elongation) (Lbs. per Sq. In.)	861	1643	714
Ultimate tensile strength (Lbs. per Sq. In.)	1966	2765	1609
Elongation (Per Cent)	64.2	50.2	68.6
Reduction in area (Per Cent)	92.4	87.9	100.0
Brinell hardness number 1 hour after casting	4.2	5.5	2.9
Brinell hardness number 10 days after casting	4.1	5.4	3.0

Elastic Limit—Faust and Tammann[48] have determined the elastic limit of lead by subjecting a highly polished specimen to a pull and observing microscopically the point at which the polish just begins to disappear. By this method the elastic limit of lead which had been melted and allowed to cool slowly was found to be 25 kilograms per square centimeter or 355 pounds per square inch.

Young's Modulus—The values found for Young's modulus for lead vary from 1500 to 1900 kilograms per square millimeter 1800 kilograms per square millimeter or 2.56×10^6 pounds per square inch may be taken as a representative value.

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BROAD ASPECTS OF THE ZINC INDUSTRY

By C. S. Trewin†

History—Zinc was probably first smelted in India or China before the Christian era. For a time it was imported into Continental Europe for the manufacture of brass. The first commercial production in Europe was at Bristol, England, in 1743. The next development, the retort method of distillation, was perfected in 1806, by Abbe Dony of Liege, Belgium. The first commercial production in the United States was in 1859, at Bethle-

†The New Jersey Zinc Company, New York City.

hem, Pa. From that time no important metallurgical advance occurred until the recent reduction to commercial practice of the electrolytic process in this country, Canada, and Australia.

The name "zinc" is attributed to the alchemist, Paracelsus, while the commercial designation "spelter" is an anglicization of the Dutch word "spiaulter" which was given to the metal by the early Dutch traders.

Occurrence—The metal occurs principally as the sulphide, blende; the carbonate, smithsonite; the silicates, willemite and calamine; and the oxides, zincite and franklinite. Zinc composes only a minute fraction of the earth's crust, but it is widely distributed geographically. The principal deposits now worked in order of their importance are found in the following countries: United States, Australia, Poland, Mexico, Germany, Italy, Canada, Spain, Algeria, French Indo-China, Burma, Peru, Sweden, Japan, Bolivia, Greece and France.

The deposits in the United States are the Appalachian group in New York, New Jersey, Virginia, and Tennessee; the Mississippi Valley group located in Wisconsin, Kansas, Missouri, and Oklahoma; and the Rocky Mountain group located in Idaho, Montana, Arizona, Utah, Colorado, and New Mexico.

Manufacture

Distillation Process—After the necessary mining, concentrating, and roasting, the ore is mixed with either anthracite or bituminous coal for charging into the distillation furnace. The furnace is a rectangular block with a longitudinal dividing wall. In both sides of this wall shelves are provided on which the rear ends of the retorts rest. The ends of the retorts resting on the shelves are slightly higher than the others which extend through the adjacent longitudinal furnace walls, forming the faces of the furnace. Several hundred retorts are placed in the two divisions of the furnace in from five to seven horizontal rows.

The retorts are made of refractory of high heat conductivity and are closed at the rear end, the walls being about one inch thick. They may be of circular or elliptical cross-section, about four feet long, with internal diameter of about eight inches. The ore and fuel is charged into the retort, after which there is placed at the retort mouth outside the furnace face a separate refractory vessel called a condenser. It has the shape of a frustrum of a cone with the butt end equivalent in size and shape to the mouth of the distillation vessel, the other end having an opening about three inches in diameter which during the distillation is luted with coal.

The furnace may be fired with natural or producer gas or coal. The products of combustion envelop the retorts, thus heating the contained charges. The reduction occurs within the retort, the metallic vapors passing into the condenser. Periodically the luting coal is removed from the condenser mouth and the liquid metal drawn into ladles and cast into commercial slab form.[1, 2, 3, 4]

Electrolytic Process—The electrolytic process consists in the main of the production of zinc sulphate solution by appropriate methods. The so-

lution is then purified of materials which would contaminate the resulting product or interfere with the efficiency of the electrolysis. The electrolysis is carried out in a cell in which generally the anode is aluminum and the cathode lead. The anodes carrying the electro-deposited zinc are periodically lifted from the cell and stripped of the metal. The metal is then melted and cast into commercial slabs.[5, 6, 7, 8]

Grades and Commercial Forms of the Metal—The recognized commercial grades in this country have been specified by the American Society for Testing Materials, and are in Table I.

Table I

Slab Zinc Specifications of American Society for Testing Materials[1]

Grade	Pb Max.	Fe Max.	Cd Max.	Al Max.	Pb+Fe +Cd Max.	Meet Specifications of
High Grade	.07	.03	.07	0	0.10	A.S.T.M. No. 1 F.S.B.[2] 91a Navy—A—(46 Z la) Army—A—(57-13 B)
Intermediate	.20	.03	.50	0	0.50	A.S.T.M. No. 2 F.S.B.[2] 91a Navy—B—(46 Z la) Army—B—(57-13 B)
Brass Special	.60	.03	.50	0	1.00	A.S.T.M. No. 3 F.S.B.[2] 91a Navy—C—(46 Z la) Army—C—(57-13 B)
Selected	.80	.04	.75	0	1.25	A.S.T.M. No. 4 F.S.B.[2] 91a Navy—D—(46 Z la) Army—D—(57-13 B)
Prime Western	1.60	.08	..	.	...	A.S.T.M. No. 5 F.S.B.[2] 91a Navy—E—(46 Z la) Army—E—(57-13 B)

[1]A.S.T.M. Standard Specifications for Slab-Zinc (Spelter) Serial Designation B 6—18, Page 524 (1927) Standards.

[2]Federal Specifications Board.

There are three domestic commercial brands which are superior to the high grade specified in the table. These are sold under brand names which designate a specific composition.

Brand names which correspond to the designations of the lower grades of the specification are used by many companies.

Commercial zinc is marketed in the form of slabs varying somewhat in dimensions, with the brand name of the producer and indentations for convenience in breaking the slab into smaller pieces cast into one surface. The slab is approximately 1½ to 2 inches thick, 9 to 10 inches wide, and 18 to 20 inches long, and weighs from 40 to 60 pounds.

The purpose for which a particular grade of zinc is to be used is fixed by its content of contaminating metals, namely, lead, cadmium, and iron.

Iron combines with zinc to form very brittle and hard constituents which in quantities above a few hundredths of a per cent seriously affect

the ductility of the metal. Cadmium is a hardening agent. Lead in quantities under one per cent exerts little effect. The presence of these metals also affects the chemical and electro-chemical properties of zinc.

In galvanizing, where the resulting coating is not to be deformed, the less pure grades of zinc are used. In cases where the galvanized coating is to be deformed, the higher grades, carrying smaller amounts of lead and cadmium are preferred.

For the manufacture of brass castings, the impurities found in commercial zinc are not of any particular moment. When brass is to be rolled and deeply drawn, the higher grade metals give superior results. In order to make brass free cutting, lead up to about 3½ per cent is added. In producing such alloys part of the lead is conveniently obtained by using the zincs of higher lead content.

When zinc is used in wet primary batteries and in electro-plating, the highest grade has been found to be commercially advantageous. For dry batteries, definite control of the impurities is required. For die casting, high grade zinc is best since the alloys generally used contain aluminum, and in such alloys the contaminating metals found in the lower grades cause serious corrosion failures.

Prices—Zinc prices in the United States are quoted on the basis of East St. Louis, Mo., for the Prime Western grade of slab zinc. The higher grades command a premium above these prices, varying according to grade from ½ to 1½ cents per pound. The following are the extreme price ranges for the past 20 years:[10]

Year		High	Low	Average
1909		6.30	4.62½	5.39
1910		6.10	4.90	5.42
1911		6.85	5.20	5.70
1912		7.50	6.25	6.93
1913		7.17½	4.95	5.61
1914		6.00	4.60	5.11
1915	Highest	27.00	5.55	14.16
1916		21.00	8.20	13.57
1917		10.87½	7.50	8.93
1918		9.50	6.62½	8.04
1919		9.00	6.00	7.04
1920		9.50	5.50	7.77
1921	Lowest	5.60	4.12½	4.67
1922		7.35	4.47½	5.74
1923		8.00	5.75	6.66
1924		7.85	5.65	6.35
1925		8.90	6.75	7.66
1926		8.75	6.70	7.37
1927		7.00	5.60	6.25
1928		6.35	5.40	6.03

Distribution of Use—The importance of zinc in our commercial structure is not generally recognized,[9] a condition which is probably due to the fact that it has been masquerading under such commercial designations as galvanizing, spelter, etc. The world's tonnage production of some non-ferrous metals, notably copper and lead, exceed it, but it is questionable if the tonnage figures represent the economic significance of the metal. This value is probably more accurately presented by consideration of the volumetric proportion of the metals produced. The volume comparison computed from the world's tonnage of zinc and copper from 1801 to 19

shows that the use of zinc was about 16 per cent greater than copper, while the 1928 world's volume of zinc was about 4.0 per cent greater than copper. This indicates that zinc is probably the most widely used non-ferrous metal.

The 1928 world production of slab zinc was 1,576,578 net tons and that for the United States was 619,595 net tons.

Principal Uses—The consumption of zinc by uses in the United States for 1928, is given in Table II.

Table II

Estimated Industrial Use of Zinc in U. S.[11]

Galvanizing (total)		291,000
Sheets	144,100	
Tubes	54,200	
Wire	41,300	
Wire Cloth	8,400	
Shapes (a)	43,000	
Brass Making		180,000
Rolled Zinc (Batteries 20,000)		73,500
Die Castings		24,000
Other Purposes (b)		58,000
		626,500

(a) includes pole-line hardware, hollow ware, chains and all articles not elsewhere mentioned.

(b) includes slab zinc used for the manufacture of French oxide, lithopone, atomized zinc dust, slush castings, and for the desilverization of lead.

The incomplete data available indicate that in other countries a greater proportion of the zinc is produced for galvanizing than is the case in the United States, and the production of rolled zinc in Europe is probably greater than in this country.

By far the greatest quantity of zinc produced is used for its corrosion resistance as indicated by the galvanizing tonnage.

The distribution for brass has not been included as it will be covered in another article.

The use of rolled zinc is principally for roofs; dry batteries; drawn and pressed objects; photo-engraving and lithographing sheets; and rolled plates used for the protection of boilers, condensers, and ship hulls from electrolytic attack.

The use of slab zinc for die castings is rapidly increasing due to economic, volumetric, and manufacturing advantages. The metal used for the manufacture of French process zinc oxide, a very high grade pigment, and for atomized zinc dust, may approach 50 per cent of the amount given in the above table for "other purposes." Slush castings consume a small tonnage.

The properties of zinc to which the wide use of the metal is chiefly due are its resistance to the atmosphere and the cheapness with which it can be applied to iron, resulting in universal application of galvanized materials; its chemical activity as evidenced by its use in batteries, for etching, and in chemical processes such as hydrogenation and precipita-

tion of heavy metals; the wide range of solid solutions formed with copper, resulting in the ductile metal brass with its many applications; its low melting point, its low volume cost, the strength of its alloys, and the ease with which they may be finished by plating, make it a superior metal for die casting.

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[7]U. C. Tainton, The Sullivan Electrolytic Zinc Plant, Engineering and Mining *Journal*, 1928, Vol. 126, p. 856.
[8]F. Laist and others, Electrolytic Zinc Plant of Anaconda Copper Mining Company at Great Falls, Montana, American Institute of Mechanical Engineers *Transactions*, 1920, Vol. 64, p. 699.
[9]U. S. Bureau of Mines, Summarized Data of Zinc Production. Washington Government Printing Office, 1929, *(Economic paper 2)*.
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[11]American Bureau of Metal Statistics. *Yearbook* 1928. N. Y., c, 1929.

CONSTITUTION OF MAGNESIUM-COPPER ALLOYS

By J. A. Gann* and A. W. Winston*

The earliest recorded work on the constitutional diagram for these alloys was by Boudouard[1] who investigated the complete system. He lo-

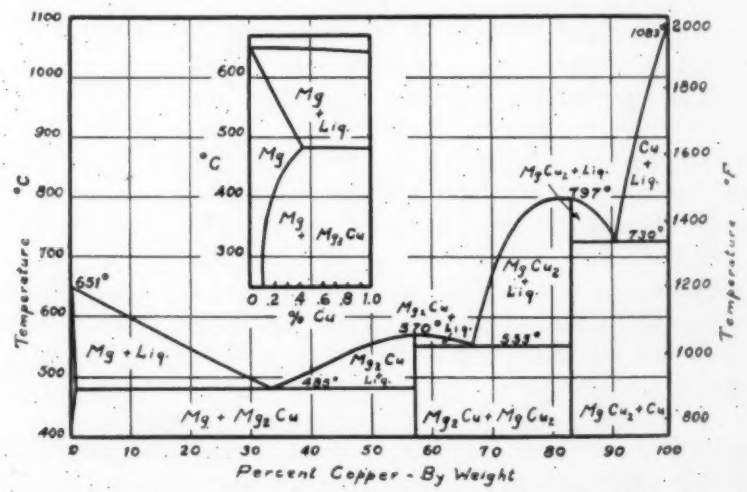

Fig. 1—Magnesium-Copper Equilibrium Diagram (Sahmen and Hansen).

cated three maxima at compositions corresponding to Cu_2Mg, $CuMg$, and $CuMg_2$, and four eutectic minima. Urazoff[2] and Sahmen[3] agreed with the general form of Boudouard's curve, but could find no evidence of the existence of the compound CuMg. More recently, Hansen[4] has published data on the magnesium end of the diagram which indicate that there is

*Metallurgists, The Dow Chemical Company, Midland, Mich.

a slight, but definite, solid solubility of copper in magnesium. This conclusion is verified by Gann.[5] The diagram in Fig. 1 is according to Sahmen with Hansen's modification inserted.

Friauf[6] has investigated the crystal structure of the compound Cu_2Mg, which he found has a face-centered cubic lattice containing eight molecules, with a value of 6.99 Å for the side of the unit cube. Runqvist, Arnfelt, and Westgren[7] recently have made a Röntgenographic examination of the magnesium-copper system. They prove the existence of two compounds only, Cu_2Mg and $CuMg_2$, thus agreeing with the thermal work of Urazoff and Sahmen. Confirming Friauf, they get from 7.021 to 7.036 Å for the parameter of the unit cube of Cu_2Mg. The second compound has a face-centered rhombic lattice, the parameters of which are: a = 5.273, b = 9.05, c = 18.21 Å, with axial ratios of 0.295 :0.497 :1, while the unit parallelepiped contains 48 atoms.

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[2]Urazoff, Russ. Physical Chemistry Society *Journal*, 1907, Vol. 39, p. 1566; Chem. Zentr., 1908, Vol. 79, p. 1038.
[3]Sahmen, *Zeitschrift für Anorganische Chemie*, 1908, Vol. 57, p. 26.
[4]Hansen, Institute of Metals *Journal*, 1927, Vol. 37, p. 93.
[5]J. A. Gann, American Institute of Mining Engineers, *Transactions*, 1929, p. 309.
[6]J. B. Friauf, American Chemical Society *Journal*, 1927, Vol. 49, p. 3107.
[7]Runqvist, Arnfelt, and Westgren, *Zeitschrift für Anorganische Chemie*, 1928, Vol. 175, p. 43.

MANGANESE AS AN ALLOYING ELEMENT IN PEARLITIC STEELS

By A. Oram Fulton†

Manganese is a very hard grayish white metal, with a reddish lustre. It takes a high polish and is not malleable. When in an extremely pure state, manganese scratches glass. The article "Physical Constants of the Principal Alloy Forming Elements" in the A. S. S. T. HANDBOOK gives the physical constants for manganese.

Historical—Manganese was first isolated in 1774 by Gahn working with Scheele. It was used as an addition to crucible steel by Heath in 1839. Mushet in 1856 patented the addition of spiegel to Bessemer steel. Mushet, also, developed the first so-called "high speed" steels which contained about 2.5 per cent manganese.

Sir Robert A. Hadfield in 1882 made a series of steels containing from 0.83 to 21.69 per cent manganese. The experiments and conclusions drawn were detailed in a paper which he presented before the British Iron and Steel Institute in 1888.[1]

Hadfield's conclusions resulted in a decided impetus being given to the manufacture and use of austenitic high manganese steels, but steel containing manganese in low percentages or pearlitic manganese steels were condemned as being brittle and having no practical applications.

[1]R. A. Hadfield, Iron & Steel Institute *Journal*, 1888, Vol. II, p. 70.

†President, Wheelock, Lovejoy & Co., Inc., Cambridge, Mass.

In the light of later developments, we now know that this erroneous conclusion with regard to low manganese steels was due to the fact that Hadfield did not work with a sufficient variety of low carbon, low manganese steels, or rather, with low manganese steels containing carbon in the proper relation to the manganese content.

In 1903 Leon Guillet reported[2] an extensive research on manganese steels. He examined two series of steels: first, the steels containing a small percentage of carbon (at the most 0.3 per cent); second, the steels in the vicinity of the eutectoid ratio containing from 0.7 to 1 per cent carbon.

The result of Guillet's experiments was to completely disprove the previously accepted theory that low manganese steels were brittle. Indeed Guillet went farther and added that "these steels possess a remarkable shock resistance, and especially a uniformity which is only met with exceptionally in carbon steels."

Following Guillet's experiments low manganese steels began to be manufactured commercially in this country and have had a steady development to the point where now they are meeting with exceeding favor by reason of their machinability, case hardening properties, strength and toughness.

Ores and their Occurrence—Manganese ores are found in many parts of the world. Their occurrence in the United States may be grouped as follows:

1. Manganese ores, containing at least 35 per cent manganese, are found most abundantly in the Appalachian and Piedmont regions of Virginia and Georgia, and on the Pacific Coast.

2. Manganiferous-iron ores, containing usually more than 5 per cent manganese and a variable proportion of iron, are found chiefly in New England, the Appalachian region, and the Lake Superior iron district.

3. Manganiferous-silver ores, consisting of mixtures of manganese, iron oxides, and hydroxides with small amounts of lead and silver, are found in the Rocky Mountain and Great Basin regions, but chiefly at Leadville, Colorado.

Enormous deposits of manganese ore are found in Southern Russia and Minas Geraes, Brazil.

Manganese itself has no uses but forms valuable alloys with copper, iron, zinc, tin, aluminum, lead, magnesium, etc. Those with iron, depending upon the percentage of manganese content, are known variously by the names spiegeleisen, ferromanganese, and silicomanganese.

Method of Adding Manganese to Steel[3]

Standard Ferromanganese—Manganese is most commonly introduced into steel in the form of ferromanganese which analyzes approximately: manganese 78 to 82 per cent, iron 15 to 19 per cent, carbon 6 to 8 per cent,

[2]Le Genie Civil, 1903, Vol. 43, p. 261-264; p. 280-282 and *Comptes Rendus,* 1903, Vol. 137, p. 480-482.

[3]W. J. Priestley, American Society for Steel Treating *Handbook* 1930.

with small percentages of silicon, phosphorus, and sulphur. Frequently, in order to lower costs, ferromanganese is added in the ladle, but a much better quality of steel is produced when the ferromanganese is added in the furnace. The latter method, however, results in a greater loss of manganese.

Low Carbon Ferromanganese—In making low carbon, low manganese steel, standard ferromanganese of the above analysis, because of its carbon content, cannot be used entirely, and must be supplemented by the use of a low carbon ferromanganese. For this purpose a low carbon ferromanganese (carbon 0.3 to 1 per cent, manganese 80 to 85 per cent) is used, which can be added either in the ladle or furnace, preferably in the furnace.

Spiegeleisen—Spiegeleisen, which in effect is a high manganese pig iron, containing 15 to 30 per cent manganese and 4.5 to 5 per cent carbon, is used in Bessemer and basic open-hearth practice in making high carbon steels, usually being added to the steel in molten form direct from a cupola.

Silicomanganese—Frequently in making low carbon steel where there is no objection to silicon, silicomanganese is used. Silicomanganese analyzes: manganese 65 to 70 per cent, silicon 16 to 25 per cent, and carbon either 1, 2 or 2.5 per cent maximum as desired. It is added either in the ladle or the furnace, preferably the latter.

Effect of Manganese on Steel—Manganese when added to steel has a beneficial effect both directly and indirectly. It increases the strength of the steel directly in that the excess manganese combines with some of the carbon to form the carbide of manganese, Mn_3C, and this carbide is found associated with the iron carbide, Fe_3C, in cementite.

It has generally been held that iron and manganese form a series of solid solutions. In the light of recent research it appears that such is not the case. The solubility of manganese in iron at room temperature seems to be close to 1.6 per cent manganese. If more than that amount is added to iron (or to steels very low in carbon) iron and manganese often display martensitic-like structures, which on examination prove to be two kinds of solid solutions. In this respect iron-manganese alloys appear to be very much like the alloys of iron and nickel.[4]

The indirect beneficial effect of adding manganese to steel is due to its action as a scavenger. It counteracts the effect of certain harmful oxides or gases and eliminates the harmful effect of sulphur by combining readily with it to form manganese sulphide, MnS.

The addition of manganese to steel lowers the transformation temperatures. A summary[5] of published results obtained in studies of pearlitic manganese steels showed that the Ac_1 point is lowered 11 degrees Fahr. for each per cent increase in manganese; the Ac_3 point is lowered 116 degrees Fahr. for each per cent increase in manganese up to 1 per cent, but only 6 to 11 degrees Fahr. for each per cent increase in manganese up

[4]Vsevolod N. Krivobok, Personal Communication.

[5]Jerome Strauss, American Society for Steel Treating TRANSACTIONS, 1928, Vol. 14, p. 2 and 3.

to 3 per cent; the Ar_2 point is lowered 125 degrees Fahr. for each per cent increase in manganese; the Ar_1 point is lowered variably, but about 90 degrees Fahr. for each per cent increase in manganese.

Manganese not only lowers the transformation points, but renders such transformation extremely sluggish. That is to say, the allotropic transformation does not take place at any definite temperature, but occupies a whole range of temperatures, sometimes as much as 150 degrees. In simpler words, it may be explained that the beginning and the end of the transformation are about 150 degrees apart, this figure depending on the amount of manganese.[6]

Guillet's constitutional diagram for manganese steel is shown in Fig. 1.

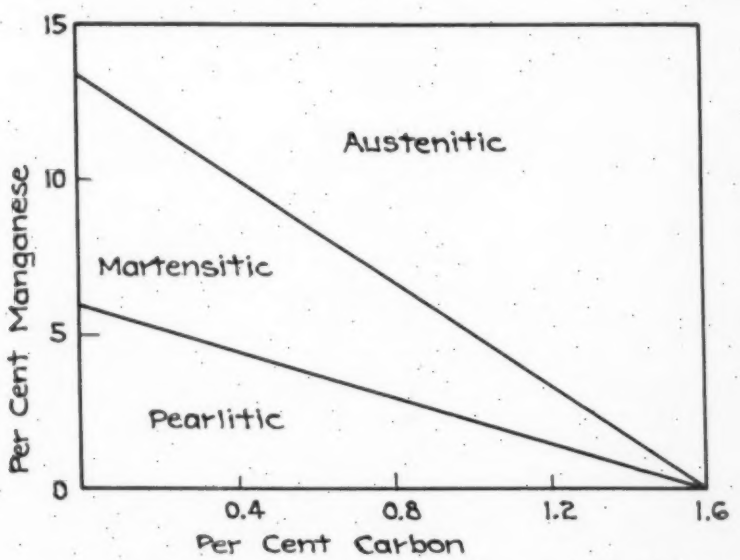


Fig. 1—Constitutional Diagram for Manganese Steel (Guillet).

It will be observed from the constitutional diagram that pearlitic manganese steels are possible with manganese content as high as 5.5 per cent provided the carbon is sufficiently low (about 0.1 per cent or under). With increasing carbon the manganese must be correspondingly lowered to maintain the pearlitic condition.

Standard Manganese Steels[7]—Table I lists the pearlitic manganese steels which are most commonly used. There are other analyses adopted by individual manufacturers and users, but the following are those having at present general distribution.

Types 2 and 5 are frequently furnished with additions of chromium in varying percentages. Types 1, 3, and 4 are also furnished with a maximum sulphur of .05 per cent.

Types 1 and 2 are most commonly used for case hardened parts, such

[6]Vsevolod N. Krivobok, Personal Communication.

[7]The use of the term "standard" here refers to standards adopted by individual manufacturers and users.

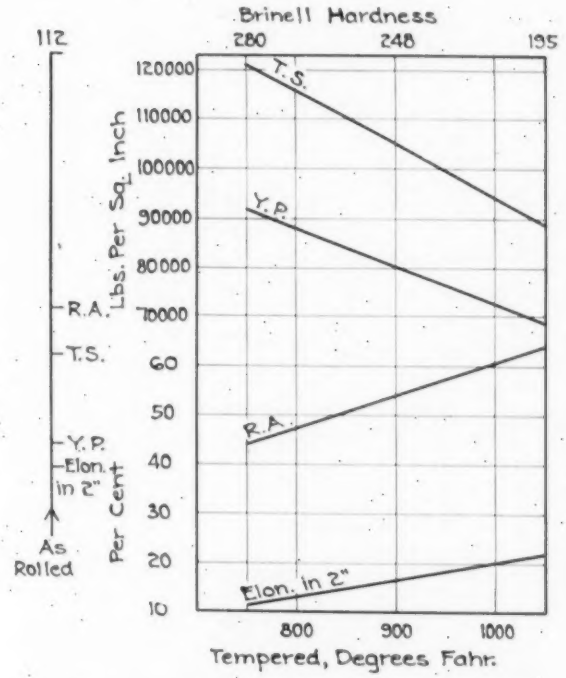


Fig. 2—Physical Properties of Manganese Steel, Type 1, Table I. C .10–.20, Mn 1.20–1.50, P .04 Max., S .075–.15. Bar ¾-inch Round Quenched into Water from 1650 Degrees Fahr.

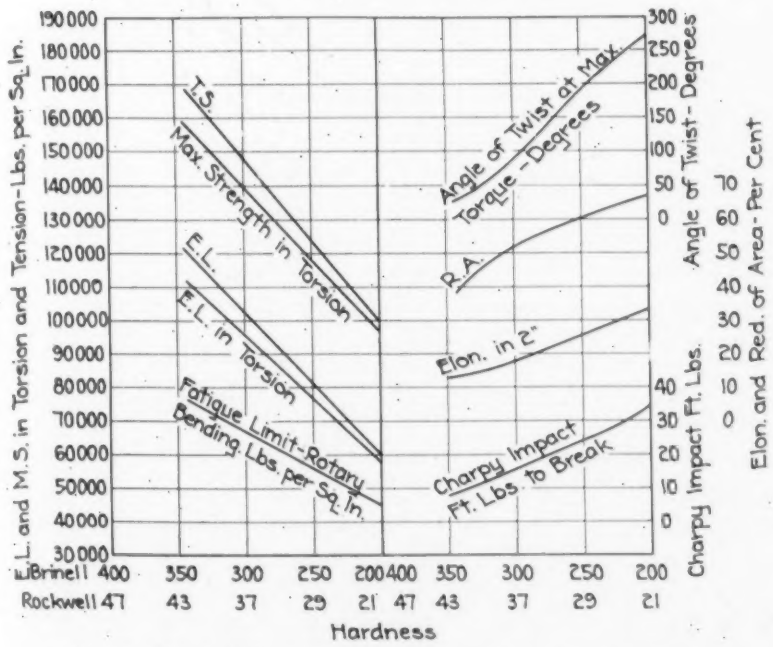


Fig. 3—Physical Properties of Manganese Steel, Type 2, Table I. C .15–.25, Mn 1.00–1.20, Cr .20, P and S .045 max. The Rockwell "C" Scale was used. Quenched into water from 1575–1600 Degrees Fahr. and Tempered to the Hardness Shown.

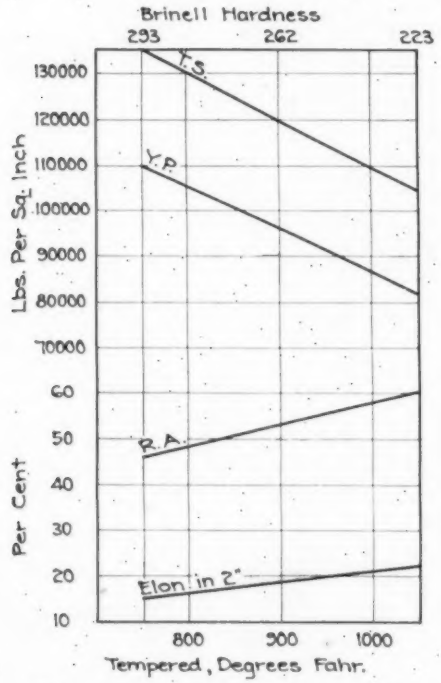


Fig. 4—Physical Properties of Manganese Steel, Type 3, Table I. C .25–.35, Mn 1.35–1.75, P .05 max., S .075–.15. Bar ¾-inch round quenched into oil from 1525 Degrees Fahr.

Table I
Pearlitic Manganese Steels

Steel Type	C	Mn	P	S
1	.10–.20	1.20–1.50	.04 Max.	.075–.15
2	.15–.25	1.00–1.20	.045 Max.	.045 Max.
3	.25–.35	1.35–1.75	.05 Max.	.075–.15
4	.30–.40	1.35–1.75	.05 Max.	.075–.15
5	.45–.55	1.10–1.30	.045 Max.	.045 Max.
1350	.45–.55	.90–1.20	.04 Max.	.055 Max.
1360	.55–.70	.90–1.20	.04 Max.	.055 Max.

as roller bearings, cam shafts, gears, etc. These steels harden with a very hard case at low temperatures and possess a strong, tough core.

The usual method of case hardening Types 1 and 2 is a double treatment after carburizing. Following is the recommended heat treatment where the double treatment reheat is used:

1. Carburize at 1600 to 1650 degrees Fahr. for sufficient time to give depth of case desired.
2. Allow work to remain in pots or boxes until cool.
3. To refine the core and give greatest toughness reheat to 1550 to 1600 degrees Fahr. and quench into oil or water (depending upon conditions, shape of piece, etc.).
4. To refine the case, again reheat to 1390 to 1425 degrees Fahr. and quench into oil, or if greater hardness is desired, heat to 1360 to 1380 degrees Fahr. and quench into water.
5. Tempering at 300 to 500 degrees Fahr. is often desirable, especially where the case is hardened in water and where grinding is to be done, as it helps prevent grinding checks.

It was pointed out above that the allotropic transformation in pearl-

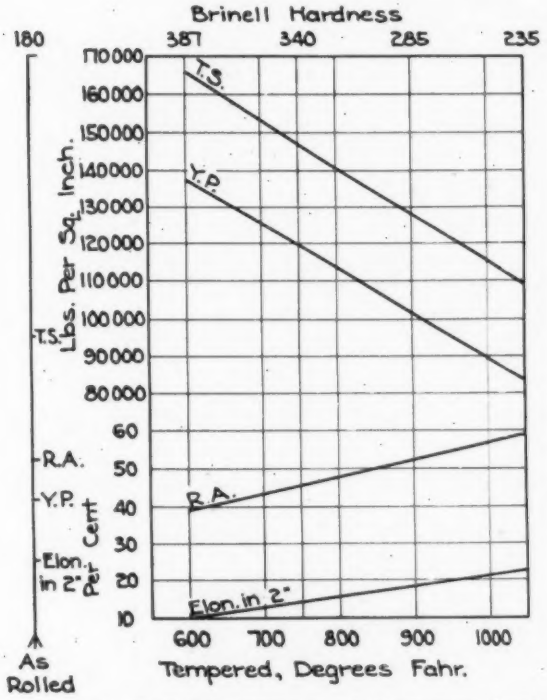


Fig. 5—Physical Properties of Manganese Steel, Type 4, Table I. C .30–.40, Mn 1.35–1.75, P .05 max., S .075–.15. Bar ¾-inch round quenched into oil from 1475 Degrees Fahr.

itic manganese steels occupies a whole range of temperatures, sometimes as much as 150 degrees apart. This means that these steels are very sluggish in passing through the critical range. It is desirable to keep the double carbides of iron and manganese in solid solution in order to prevent brittleness. With Types 1 and 2 this can be done best by quenching them direct from the pot rather than to permit the carbides to be precipitated from solid solution by cooling in the pot. When this direct quench from the pot is used, it should, of course, be followed by subsequent treatment for the refinement of the case.

Types 1 and 2 are frequently heat treated for strength and toughness by direct quench in water without carburizing and then reheating as required. Excellent physical properties result. Figs. 2 and 3 show the physical properties of these two steels with varying tempering temperatures or at different conditions of hardness obtained by varying heat treatments.

Types 3 and 4 are used where higher strength and toughness are required in combination with good machining qualities. Common uses are heat treated shafts, gears, spindles, etc. The physical properties of these grades are shown in Figs. 4 and 5. The heat treatments recommended for any given combination of hardness, strength, and toughness within the limitations of the steels are given in the Figures.

Steel Type 5 is used where very high physical properties are re-

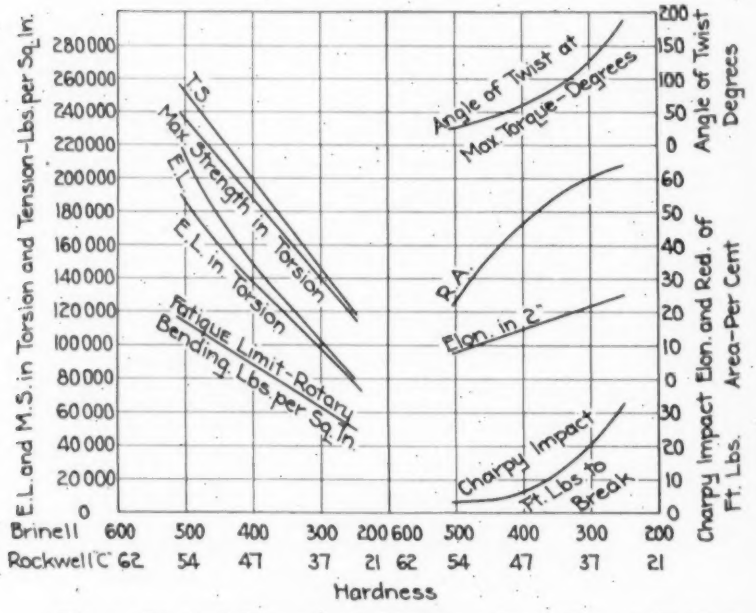


Fig. 6—Physical Property of Manganese Steel, Type 5, Table I. C .45–.55, Mn 1.10–1.30, Cr approx. .50, P and S .045 max. Quenched into oil from 1475–1500 Degrees Fahr. and Tempered to the Hardness Shown.

quired. Common uses are gears, shafts, clutches, bolts, etc. This type is invariably quenched into oil. The physical properties of this steel are shown in Fig. 6 when heat treated to the varying degrees of hardness indicated in the abscissae.

The heat treatment consists of quenching into oil from 1475 to 1500 degrees Fahr. followed by tempering to the hardness desired.

S.A.E. Steels Nos. 1350 and 1360 are intended for helical springs made of round, cold-drawn wire up to $\frac{3}{16}$-inch diameter. Steel 1350 is recommended for wire from 0.025-0.100-inch diameter and steel 1360 for wire 0.100 inch diameter and larger.

General Properties of Pearlitic Manganese Steels

Welding—These steels cannot properly be classed as "welding steels," but may be used without difficulty in the manufacture of parts which involve welding in their fabrication.

Forging—None of these steels present any unusual difficulties in any of the regular forging operations, and can be freely used in parts which require forging as a necessary preliminary to their manufacture.

Cold Drawing—Pearlitic manganese steels may be cold drawn without difficulty. Like other alloy steels, it is recommended when cold drawing steels of .35 per cent carbon or more that they be annealed before the drawing operation.

Machining—As a class, the pearlitic manganese steels are particularly free machining. Indeed, the low carbon steels, such as Types 1 and 2 in the Table I are practically equivalent in their machining properties to

Bessemer screw stock. As the carbon, manganese, and any other alloying elements are increased, the machining becomes more difficult. In the case of steels of over .40 per cent carbon, it is recommended that they be annealed before machining.

Other Types of Manganese Steels—Pearlitic manganese steels are made with the manganese content higher (up to 3 per cent) than those steels described above, but have a limited and special use. Steels are made containing manganese from 1 to 2 per cent in combination in varying percentages with nickel, chromium, molybdenum, silicon and other elements. These steels likewise are limited in use and have a highly specialized application. While they possess unusually good physical properties, they have presented too many manufacturing difficulties to the user for general adoption. The chief difficulties are that they add considerable cost in fabrication due to difficulty in the forging and machining operations.

Manganese in Tool Steel—The average "normal" manganese content in tool steel is 0.20 to 0.35 per cent. One special type of manganese tool steel has very decided non-changeable or non-deforming qualities, however, and is very generally used. The composition of this steel varies somewhat with different manufacturers. The approximate chemical composition is as follows;

Carbon	.90 to 1.00 per cent
Manganese	1.00 to 1.50 per cent

Some manufacturers run the manganese as high as 1.80 per cent, with or without vanadium, while others add 0.50 per cent chromium, 0.50 per cent tungsten, or about 0.25 per cent vanadium, or combinations of these. This steel, however, depends essentially for its qualities upon the carbon and manganese content.

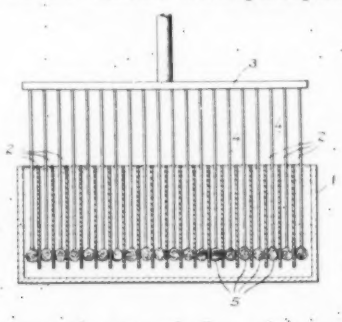
The heat treatment consists in quenching in oil from approximately 1450 degrees Fahr. followed by tempering as required, usually 325 to 400 degrees Fahr.

Reviews of Recent Patents

By NELSON LITTELL, Patent Attorney
475 Fifth Ave., New York City—Member of A. S. S. T.

Re. 17,484, Nov. 12, 1929, Rustproofing Process, William H. Allen, of Detroit, Michigan, Assignor, By Mesne Assignments, to Parker Rust Proof Company, of Detroit, Michigan, a corporation of Michigan.

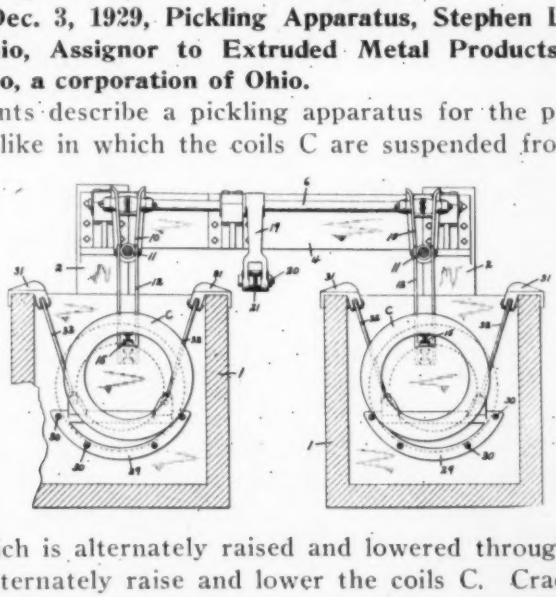
This patent describes a rustproofing process in which the sheets 2 are immersed in the tank 1 to be coated with phosphate for rustproofing and



are rubbed to prevent crystal growth by alternate raising and lowering of the frame 3 which carries a series of pads 5 on the end of the rods 4 which pads rub against the surface of the plates 2.

1,738,412, Dec. 3, 1929, Pickling Apparatus, Stephen L. Williams, of Bridgeport, Ohio, Assignor to Extruded Metal Products Company, of Bridgeport, Ohio, a corporation of Ohio.

These patents describe a pickling apparatus for the pickling of coils of wire or the like in which the coils C are suspended from a horizontal

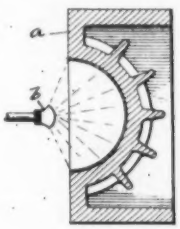


member 15 which is alternately raised and lowered through oscillation of the rod 6 to alternately raise and lower the coils C. Cradles 29 are immersed in the pickling baths by means of links 32 and are located eccentrically with reference to the horizontal member 15, so that as the coils C are lowered into the cradle 29, they are given a rotary movement by the shape of the cradle. In the patent No. 1,734,986 the cradles are auto-

matically moved in synchrony with the movement of the horizontal member.

1,735,120, Nov. 12, 1929, Mold Coating, Harry S. Lee, of Plymouth, Michigan.

This patent shows a machine for continuously coating the surface of permanent molds *a* by spraying the face thereof with a mixture of lamp



black and graphite carried in kerosene from the nozzle *b*. From the molding machine, the molds are carried on a rotating turret and sprayed at a certain point in their rotation by suitably located nozzles *b*.

1,735,842, Nov. 19, 1929, Process for Rustproofing Articles of Iron and Steel, William H. Allen, of Detroit, Michigan, Assignor, by Mesne Assignments, to Parker Rust Proof Company of Detroit, Michigan.

This patent describes a process of rustproofing in which it is unnecessary to heat the rustproofing solution containing the articles to be coated to boiling or to accurately control the conditions of the process. According to the process of this patent, the articles may be dipped cold or moderately warm into the rustproofing solution, allowed to remain a short period of time and after removal are heated to a temperature of approximately 450 degrees Fahr. The solution should contain about 2½ per cent phosphoric acid and about 0.175 per cent of chromium combined as phosphate and the subsequent heating of the articles is to convert the chromium phosphate into an insoluble chromium pyro-phosphate.

1,736,120, Nov. 19, 1929, Molybdenum Alloy Iron and Steel, Alan Kissock, of Forest Hills, New York.

This patent describes an alloy for use in alloying molybdenum with steel or iron which comprises a molybdenum-bearing pig iron produced by adding a molybdenum compound, such as molybdenum oxide or calcium molybdate in the blast furnace, so that it is reduced in the blast furnace operation and removed from the blast furnace as a molybdenum-bearing pig iron. The molybdenum-bearing pig iron can, if desired, be added to the charge in the steel-making furnace while still molten.

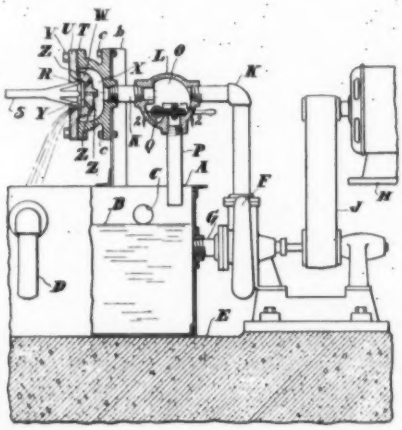
1,736,457, Nov. 19, 1929, Composition of Matter for and Method of Purifying Fused Salt Baths, William J. Merten, of Pittsburgh, and Charles T. Gayley, of Swissvale, Pennsylvania, Assignors to Westinghouse Electric and Manufacturing Company, a corporation of Pennsylvania.

This patent describes a material for eliminating sulphides, oxides and other impurities from fused salt baths used for the heat treatment of steel. The preferred material consists of boric acid mixed with 10 to 25 per cent

of charcoal or other carbonaceous material. The boric acid or carbon reacts with the impurities of the bath and removes the impurities as gases or as the scum or dross on the surface of the bath.

1,738,050, Dec. 3, 1929, Quenching Device for Drill Steels, Charles C. Hansen, of Easton, Pennsylvania, Assignor to Ingersoll-Rand Company, of Jersey City, New Jersey, a corporation of New Jersey.

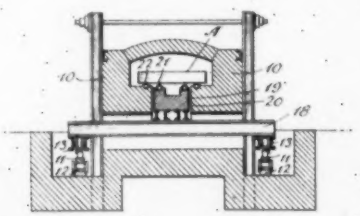
This patent describes a device for quenching the points of drill bits so as to prevent the formation of air bubbles on the heated surface and



unequal quenching which comprises inserting the point of the drill bit R in a socket member W provided with a plurality of converging apertures Z through which the quenching fluid flows from the chamber X to spray the end of the drill bit under considerable pressure. The quenching fluid is pumped by means of a continuous pump F from the tank B to the chamber X and when the chamber X is not used for the quenching of drill bits, the fluid may be permitted to return to the tank B through the pipe P by suitable manipulation of the valve Q.

1,739,250, Dec. 10, 1929, Furnace, Fritz Menne, of Weidenau-on-the-Sieg, Germany, Assignor to Herman F. Hoevel, of New York, N. Y.

This patent describes an improvement in continuous furnaces in which the material passing therethrough is alternately raised, moved forward



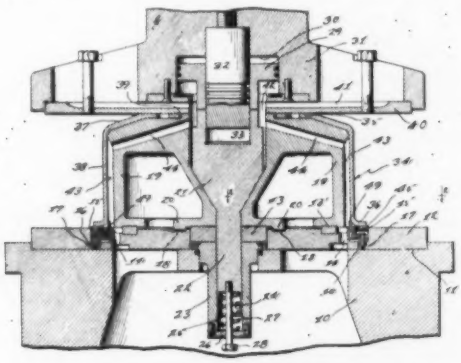
and deposited by reciprocating members in which the articles, in passing through the furnace, are supported on cooling pipes 22 and the oscillating article moving member 19 is also provided with cooling pipes 21 during its support of the articles.

1,736,919, Nov. 26, 1929, Surface-Hardened Material and Method for Producing the Same, Augustus B. Kinzel, of New York, N. Y., Assignor to Electro Metallurgical Company, a corporation of West Virginia.

This patent describes a process of surface hardening metal articles by nitriding in which the article is first coated with aluminum by heating the steel article in contact with aluminum powder under neutral or reducing conditions at a temperature of about 500 to 800 degrees Cent. for a period of about two hours and then heating the surface-aluminized steel to about 500 to 580 degrees Cent. in the presence of ammonia to produce a hard nitrogenized compound on the surface only of the steel article without detrimentally affecting the core of the article.

1,740,347, Dec. 17, 1929, Method of Hardening Articles, Walter G. Hildorf, of Lansing, Michigan, Assignor to Reo Motor Car Company, of Lansing, Michigan, a corporation of Michigan.

This patent describes a mechanism for the differential quenching of articles, such as brake drums 38, in which the brake drums are clamped



between suitable supports 19 and 40 and immersed in a quenching fluid, such as oil, to force the oil upwardly through passages 43, 44, along the inside of the brake drum before the oil contacts with the outer surface 34 of the brake drum to cause a greater degree of quenching and hardness on the interior of the brake drum.

1,737,255, Nov. 26, 1929, Hard Alloy and Process of Manufacturing the Same, Philip M. McKenna, of Latrobe, Pennsylvania, Assignor to Vanadium Alloys Steel Company, of Latrobe, Pennsylvania.

This patent describes a hard alloy having sufficient strength to withstand mechanical shock for use as a lathe-tool milling cutter and the like. The alloy consists of tungsten carbide (containing 2.82 per cent carbon) eighty parts and beryllium four and one-half parts. The alloy is produced by carburizing the tungsten and adding the beryllium metal in comminuted condition with the tungsten carbide, pressing the powdered mixture into a steel mold at a pressure of 45,000 pounds per square inch, then inserting the pressed body in a graphite receptacle and heating the same to a temperature of about 2350 degrees Fahr. for approximately seven hours. When cooled, the sintered body may be shaped by grinding and then heated to about its melting point to produce a hard alloy.

THE ENGINEERING INDEX

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In the preparation of the Engineering Index by the staff of the American Society of Mechanical Engineers some 1700 domestic and foreign technical publications received by the Engineering Societies Library (New York) are regularly searched for articles giving the results of the world's most recent engineering and scientific research, thought, and experience. From this wealth of material the A. S. S. T. is supplied with this selective index to those articles which deal particularly with steel treating and related subjects.

AIRPLANE ENGINES

ALUMINUM. Modern Light Alloys and Their Application to Aircraft-Engine Design, G. D. Welty. *Soc. Automotive Engrs.—Jl.*, vol. 25, no. 5, Nov. 1929, pp. 469-473 and (discussion) 473 and 488, 1 fig.

More important commercial aluminum-base alloys are discussed; main physical characteristics; methods of fabrication; alloys best suited to various aircraft-engine requirements; physical properties of various casting alloys at elevated temperatures; special purpose alloys; new aluminum alloy for pistons; application of magnesium alloys to aircraft-engine design; thorough tests recommended before adopting magnesium for major engine parts.

AIRPLANE FUSELAGES

MANUFACTURE. Chromium-Molybdenum Steel Tubing Fuselage Construction, J. H. Kindelberger. *Soc. Automotive Engrs.—Jl.*, vol. 25, no. 5, Nov. 1929, pp. 474-477, 6 figs.

Reason why pilots, operators, maintenance men, designers, and manufacturers prefer welded steel to stick and wire construction for airplanes; molybdenum steel developed to make stronger struts possible; strength and reliability of welds depends on designing joints that can be welded without overheating any parts; complicated fittings should be made up into subassemblies; jigs for welding airplane assemblies, including fixtures for drilling holes in assembled frame; measures taken to prevent corrosion.

AIRPLANE MANUFACTURE

METAL CONSTRUCTION. The Uses of Metal. *Aircraft Eng. (Lond.)*, vol. 1, no. 7, Sept. 1929, pp. 217-218.

Importance of metal construction for airplanes and its prospects for future of general engineering industries are discussed; airplane has at last emerged from amateur stage and very definitely become engineer's job; present use of metal as covering particularly for wings is mistaken; airplane of tomorrow will depend on metal skin for considerable portion of its strength.

AIRPLANE MATERIALS

ALUMINUM ALLOYS. Wrought Alloys of Aluminum in Aircraft, G. Mortimer. *Aircraft Eng. (Lond.)*, vol. 1, no. 7, Sept. 1929, pp. 223-227, 9 figs.

Survey of various classes of aluminum alloys, with detailed considerations of their properties and methods of treating them; magnesium silicide; copper eutectic compound; combination of two media; duralumin; annealing; heat treatment methods; corrosion; influence of design; metallurgical precautions; oxide protection of metals; organic coatings; electrodeposition.

ALUMINUM ALLOYS. Service Characteristics of Light Alloys, E. H. Dix, Jr. *Soc. Automotive Engrs.—Jl.*, vol. 25, no. 5, Nov. 1929, pp. 463-468 and (discussion) 468.

Discussion of aluminum and magnesium, in their pure and alloyed states, as most suitable for aircraft construction; physical properties; effects of adding small quantities of alloying elements; heat treating for increasing strength per unit weight of alloys; effect of natural aging and artificial aging at elevated temperatures and of quenching in hot and cold water after heat treating; several types of corrosion; protection afforded to aluminum alloys by surface coating of pure aluminum.

MAGNESIUM ALLOYS. Magnesium Alloys in Aeronautics. *Airway Age*, vol. 10, no. 12, Dec. 1929, pp. 1937-1939, 9 figs.

Many applications of magnesium alloys, casting and forgings in airplane industry are discussed; magnesium castings offer combination of high strength and light weight; properties of AM7.4 alloys in as cast and heat treated state; for rolled sheet, plate or bar, pure magnesium or alloy containing 4 per cent of aluminum and 0.4 per cent of manganese are used; precautions to be observed in casting magnesium; uses in engine and plane.

AIRSHIP MANUFACTURE

METAL CONSTRUCTION. Heat Treats Stainless Steel Tubes. *Iron Age*, vol. 125, no. 18, Oct. 31, 1929, pp. 1157-1160, 6 figs.

Description of new type of members, materials and methods of manufacture developed by Boulton and Paul, Norwich, England, for manufacturing giant airship R-101; soft strip is formed into tubing and drawn through pair of dies with electric furnace between; plant for fabricating shapes; physical properties of steel used; continuous heat treatment process; uniformity of product; control tests on heat treated sections.

ALLOY STEEL

HEAT-RESISTING—PETROLEUM INDUSTRY. The Use of Steel in the Oil Industry with a Special Reference to Heat-Resisting Steels, R. Hadfield and S. A. Main. *Instn. of Petroleum Technologists*

—Jl. (Lond.), vol. 15, no. 76, Oct. 1929, pp. 611-633, 13 figs. on supp. plates.

Importance of using best steel in oil-well drilling and pumping; qualities; forging; heat treatment; alloy steel; review of processes of manufacturing heat-resisting and non-corrodible steels.

ALLOY STEEL PROPERTIES

Study of Special Steels (Etude documentaire, descriptive et technique sur les aciers spéciaux). *Revue Industrielle (Paris)*, vol. 59, no. 2242, Sept. 1929, pp. 620-629, 9 figs.

Yearly production; superiority of special over carbon steels; manufacturing methods; general properties; classification; influence of alloying elements; chemical, physical, and electrical properties; various characteristic curves are given.

ALLOYS

EQUILIBRIUM DIAGRAMS. Solidus Determination by Means of Heating Curves, J. H. Andrew and H. M'Neil. *Metallurgist (Supp. to Engineer, Lond.)*, Oct. 25, 1929, pp. 154-155, 1 fig.

Review of paper in Jl. of Roy. Tech. College; authors suggest that, with regard to certain alloys freezing as solid solutions, solidus may not be separated from liquidus by such wide range of temperature as that depicted in accepted thermal equilibrium diagrams; this suggestion is based upon fact that normally cooled solid solutions are usually cored and not homogeneous; for purpose of examining this hypothesis authors selected copper-tin alloy containing 7.89 per cent tin.

MAGNETIC. Alloys with High Permeability (Les alliages à haute permeabilité), Chaughat. *Société Francaise des Electriciens—Bul. (Paris)*, vol. 9, no. 98, Oct. 1929, pp. 1091-1106, 7 figs.

Difficulties of using ordinary magnetic materials in telephone transformers; ferromagnetic materials of high permeability; classification and properties of ferronickel; manufacture of nickel alloys; applications in telephone industry; examples of calculations.

SURFACE TENSION. Surface Tension of Gold, Zinc, Gold-Copper, Silver-Copper and Iron Alloys (Die Oberflaechenspannung von Gold, Zink, Gold-Kupfer-, Silber-Kupfer- und Eisenlegierungen), W. Krause and F. Sauerwald. *Zeit. fuer Anorganische und Allgemeine Chemie (Leipzig)*, vol. 181, no. 4, July 20, 1929, pp. 353-371, 6 figs.

Values for surface tension found were higher in all cases than those recorded in literature; for alloys of silver and copper surface tension decreases linearly up to 64 per cent copper, above which it increases linearly with temperature; measurements were made from 510-640 deg.; iron alloys show great difference in surface tension.

ALUMINUM ALLOYS

PROTECTIVE COATINGS. Some Recent Developments in Protective Coatings for Aluminum and Its Alloys, H. Sutton. *Metal Industry (Lond.)*, vol. 35, no. 18, Nov. 1, 1929, pp. 411-412 and (discussion) 412-414.

Brief survey of recent developments in production of protective coatings on metals; anodic oxidation of aluminum and its alloys; chlorides undesirable; scope of anodic process; Jirotka process; Stafford O'Brien process; anodic treatment of steel; application of pure aluminum; coatings of metallic aluminum; electrodeposits on aluminum.

ALUMINUM BRONZE

On the Nature of Eutectoid-Transformation in Relation to the Mechanism of Quenching and Tempering of Aluminum-Bronze, I. Obinata. *Ryojun College of Eng.—Memoirs (Ryojun)*, vol. 2, no. 3, Aug. 1929, pp. 205-225, 11 figs.

Experiments covering preparation of specimens; measurement of electric resistivity, including change in resistance of annealed and quenched specimens caused by change of temperature and of quenched specimens due to soaking at various temperatures; thermal analyses; microscopic investigations; theoretical considerations on nature of eutectoid-transformation. (In English.)

ALUMINUM CORROSION

Influence of Different Salts on the Solution of Pure Aluminum in Hydrochloric Acid (Influence de divers sels sur la dissolution de l'Aluminium pur dans l'acide chlorhydrique), J. Calvet. *Académie des Sciences—Comptes Rendus (Paris)*, vol. 189, no. 4, July 22, 1929, pp. 183-186.

Author continues description of tests on solution of extra-pure aluminum in dilute solutions of hydrochloric acid, and examines effect produced on its long period of passivity by addition of small quantities of salts of other metals; table gives concentrations of different salts per liter of solution, duration of tests, and loss in weight of aluminum specimens in mg. per sq. cm.

ALUMINUM—SILICON SOLUBILITY

Aluminum and Its Formation of Mixed Crystals with Silicon (Aluminium und seine Mischkristallbildung mit Silicium), W. Koester. *Zeit. fuer Anorganische und Allgemeine Chemie (Leipzig)*, vol. 181, no. 3, July 6, 1929, pp. 295-297, 1 fig.

Remarks pertaining to publication under same title by L. Anastasiadis; criticizes technic employed and results obtained by author in determination of temperature coefficient of solubility of silicon in aluminum and in measuring electric resistance of aluminum.

AUTOMOBILE AXLES

HEAT TREATMENT. Salisbury Lowers Hardening Costs by Using Electric Furnaces, W. S. Scott. *Automotive Industries*, vol. 61, no. 19, Nov. 9, 1929, pp. 681-683, 3 figs.

Description of method employed by Salisbury Axle Co., Jamestown, N. Y., in manufacture of automobile axles; savings in operating time and maintenance expense are made by installation of two continuous-pusher electric furnaces which replace four oil-fired box heaters; alloy steel used in axles; operating cost and fuel and electricity costs; comparative costs; labor for heat treatment; saving in labor after heat treatment.

AUTOMOBILE ENGINES

CRANKSHAFTS — HEAT TREATMENT. Heat Treatment of Crankshaft Dies and Forgings, C. B. Phillips. *Fuels and Furnaces*, vol. 7, no. 2, Nov. 1929, pp. 1761-1763, 3 figs.

Crankshafts, after finish forging and final upset, are heated in continuous, automatic, oil-fired normalizing furnace at temperature of 1650 deg. fahr. and then quenched, eliminating all subsequent heating operations; dies are heated in gas-fired furnace to temperature 1500 deg. fahr. quenched and drawn at temperature of 700-1000 deg. fahr. See *Engineering Index*, 1928, p. 178.

AUTOMOBILE PLANTS

HEAT TREATING FURNACES. Automatic Heat-Treating Unit, W. M. Hepburn. *Iron Age*, vol. 124, no. 20, Nov. 14, 1929, pp. 1304-1306, 4 figs.

Description of three interconnected furnaces installed at Studebaker Plant for purpose of normalizing, hardening, quenching and drawing; these functions may be separated for various heat treatments, according to requirements of production; furnaces are fired with both gas and oil, gas burners being used in soaking-zone; waste-heat recuperator.

BEARING METALS

Bearing Alloys, R. T. Rolfe. *Mech. World (Lond.)*, vol. 86, no. 2236, Nov. 8, 1929, pp. 440-441, 5 figs.

Question of lubrication and bearing metals are discussed; theory of lubrication is described and illustrated; behavior of different lubricating oils; requirements of bearing metals; examples of hard bearing metals. Abstract of paper presented before Manchester Assn. of Engrs. (To be continued.)

A New Product of Bearing-Metal Research. *Metal Industry (Lond.)*, vol. 35, no. 19, Nov. 8, 1929, pp. 441-442.

Description of alloy bearing metal known as Thermit or Universal Bearing Metal, which had been produced by T. Goldschmidt, A.G., Essen, in connection with G. von Hanffstengel; structure of Thermit comprises uniformly distributed hardening bodies bedded in plastic eutectic ground-mass which is rich in lead; through high softening point of Thermit alloys, favorable hardness conditions are maintained; slipperiness of Thermit investigated by process of von Hanffstengel; casting of Thermit. Abstracted from Technische Blaetter, Aug. 4.

BERYLLIUM

Beryllium and Beryl, A. V. Peter. *U. S. Bur. of Mines—Information Cir.*, no. 6190, Nov. 1929, 20 pp.

General notes; description, properties, and uses of beryllium; historical data; prices and market conditions; beryllium minerals; description, uses, and occurrences of beryl; location of beryl deposits, by states in United States and countries; minor minerals; beryl market and prices. Bibliography.

BOILER CORROSION

Decomposition of Efflorescence on Leaking Boilers Due to Flue Gases (Zersetzung von Ausbluehungen an undichten Kesseln durch die Rauchgase). *Zeit. des Bayerischen Revisions-Vereins (Munich)*, vol. 33, no. 20, Oct. 31, 1929, pp. 279-280.

Results of investigation of efflorescence on boiler supplied with feedwater softened according to lime-soda process; it is shown that flue gases have disintegrating effect on such incrustations.

BRASS CORROSION

The Corrosion of Brass, M. Haas. *Metals and Alloys*, vol. 1, no. 4, Oct. 1929, pp. 183-189, 12 figs.

Discussion of relations between corrosion and structure especially with problem of much-discussed grain size; results of various investigators dealing with relations between grain and corrosion; investigations of typical corrosion appearances on automobile radiator sheets; pronounced corrosion occurred with non-uniformly coarse grain in case of 64 and 63 brass; suggestions to avoid corrosion in brass sheets for radiators. Bibliography. Article translated from Korrosion und Metallschutz, Feb. 1929, previously indexed.

BRASS PROPERTIES

Properties of Hard Brass (Die Eigenschaften des Hartmessings MS 58), O. Bauer and K. Memmler. Mitteilungen der deutschen Materialpruefungsanstalten, no. 8. Berlin, Julius Springer, 1929, 76 figs. tables and diagrams, 13.50 R.M.

Properties of different standard brass varieties were investigated at request of German Industrial Standards Committee by Staal. Materialpruefungsamt and Kaiser-Wilhelm Institut fuer Metallforschung in Berlin-Dahlem; material was supplied by different firms made by different processes, and are to some extent of different chemical composition (varying lead content); methods and results of tests are described.

BRAZING

Copper Brazing in Electric Furnaces, H. M. Webber. *Machy. (N.Y.)*, vol. 36, no. 3, Nov. 1929, pp. 205-207, 3 figs.

Fabricating drawn and cast steel parts by copper brazing in electric furnace with hydrogen atmosphere is described; temperatures used are around 2100 deg. fahr. which is slightly above melting point of copper; steel is annealed in process by slow cooling in reducing atmosphere; less pickling or cleaning is required; applications of methods in copper brazing electric-refrigerator parts, such as evaporators, floats, check valves, and pistons.

BRONZES, HIGH STRENGTH

Notes on Bronze for High Strength (Notes sur les bronzes à hautes résistances), L. Achard. *Fonderie Moderne (Paris)*, vol. 23, Oct. 25, 1929, pp. 513-514.

Composition and properties of high-strength bronzes and aluminum are discussed.

CASE HARDENING

Comparative Study of Heat Treatment of Case-Hardened Steel (Vergleichende Untersuchungen ueber die Waermebehandlung eingesetzter Staehle), H. Luepfert. *Stahl und Eisen (Duesseldorf)*, vol. 49, no. 48, Nov. 28, 1929, pp. 1717-1724, 17 figs.

Relation of grain size to annealing time and temperature and quenching speed of carbon and chromium-nickel steel; comparison of different heat treating processes based on bending tests and structure analysis; rapid quenching after case hardening followed by brief intermediary cooling and annealing at 200 deg. cent. is recommended.

CAST IRON RESEARCH

German Foundry Research Work, Piwowarsky. *Foundry Trade Jl. (Lond.)*, vol. 41, no. 690, Nov. 7, 1929, p. 333, 2 figs.

Research being carried out under author's supervision is outlined; growth of cast iron studied; researches on degasified cast irons; work on machinability not completed; researches on influence of oxygen on cast iron; working and conduct of cupola carefully studied; question of carbon pick-up in cupola; steel castings.

Rejuvenation of Gray Iron Planned, L. W. Spring. *Iron Age*, vol. 124, no. 20, Nov. 14, 1929, pp. 1321-1322.

Properties of cast iron and some of their disadvantages are discussed, and various types of cast iron, including soft or gray cast iron, semi-steel or high-test cast iron, chilling irons, malleable cast iron, and pearlitic irons are described.

CAST IRON TEST BARS

Test Bars Vary with Size of Castings in New Standards, W. R. Needham. *Iron Age*, vol. 124, no. 22, Nov. 28, 1929, p. 1438.

Discussion of new British Engineering Standards issued covering test pieces for gray iron castings; research has made it quite clear that average tensile strength of gray iron decreases thicker sections, since rate of cooling in mold profoundly affects solidified metals; for transverse tests, too, three sizes of circular bar are now made to serve for these three casting thickness ranges.

CAST IRON TESTING

French Foundry Congress. *Engineer (Lond.)*, vol. 148, no. 3851, Nov. 1, 1929, pp. 475-476.

Congress organized by Association Technique de Fonderie and held in October at Ecole Nationale des Arts et Métiers in Paris, gave special prominence to methods of testing cast iron; Le Thomas read paper in collaboration with R. Bois on methods of testing cast iron; paper that provoked long discussion was that on production of high-resistance gray cast iron in electric furnace, by E. Valenta, of Skoda Works; communication on cast-iron parts for agricultural machines was received from H. Bornstein of Moline, Ill.

CASTINGS

TESTING. Modern Methods of Testing Castings. *Foundry Trade Jl., (Lond.)*, vol. 41, no. 690, Nov. 7, 1929, p. 342.

Modern methods of testing and examination and their influence on design of castings are discussed; optical methods of investigating sectional models of casting, using elastic stress determination advocated by Coker; separately cast test bars; electromagnetic systems of testing; X-ray examination of casting.

X-RAY ANALYSIS. X-Ray Inspection of Castings. *Soc. Automotive Engrs.—Jl.*, vol. 25, no. 5, Nov. 1929, p. 551.

Value of X-ray analysis of castings is discussed; its solution to problem of over-zealous and careless inspectors; X-ray inspection used successfully in examination of cast steel, bronze, and aluminum as well as other articles.

CHEMICAL EQUIPMENT

STAINLESS STEEL. The Production of Chemical Plant of Acid-Resisting Steel, W. H. Hatfield. *Engineering (Lond.)*, vol. 128, no. 3329, Nov. 1, 1929, p. 567.

Range of steel considered covers chromium content of 14 to 20 per cent, nickel content of 7 to 12 per cent, and carbon not exceeding 0.20; welding plays important part in production of chemical plant from these steels; hot riveting has been successfully conducted on extensive scale; steels can be made into castings of all shapes and dimensions; consideration of instances where trouble has been experienced. Abstract of lecture before Instn. Chem. Engrs.

CHROMIUM-NICKEL STEEL

Technological and Metallographic Properties of Highly Alloyed Chromium-Nickel Steel (Technologische und metallographische Eigenschaften eines hochlegierten Chrom-Nickel-Stahles), E. Greulich and G. Bedeschi. *Archiv. fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 3, no. 5, Nov. 1929, pp. 359-363, 15 figs.

Influence of heat treatment on engineering properties of cold-rolled austenitic chromium-nickel steel; grain size in recrystallization zone; separation of carbides and their redissolution at higher temperatures; influence of toughness; with almost complete dissolution at 1100 to 1150 deg., highest degree of toughness is reached within this narrow temperature range.

Permanence. *Metallurgist (Supp. to Engineer, Lond.)*, Nov. 29, 1929, pp. 161-162.

Comments on view put forward by one of most important metallurgical concerns on Continent, which manufactures nickel-chromium steel, that these steels especially those of type used in heat-treated state for structural purposes are not permanently stable or reliable; phenomena of temper brittleness is discussed; question suggests itself whether changes which have been observed in nickel-chromium steel after lapse of time may not be changes which may be described as development, by aging, of temper brittleness latent in steel and only suppressed by quenching after tempering.

COPPER ALLOYS

Hardened Copper for Electrical and Chemical Equipment, W. N. Hibbard. *Iron Age*, vol. 124, no. 22, Nov. 28, 1929, p. 1434.

Uses and properties of hardened coppers known as Everdur and Tempaloy and their welding characteristics are described; table shows composition and physical properties of soft and hardened copper; hardened copper alloys are strong or stronger than medium carbon steels; cold-rolled alloys weld easily and will not season or corrosion crack. Ab-

stract of paper presented before Int. Acetylene Assn.

MELTING. Development of New Deoxidation and Melting Process for Metals, Especially Copper Alloys (Die Entwicklung eines neuen Deoxydations und Schmelzverfahrens fuer Metalle insbesondere fuer Kupferlegierungen), W. Reitmeister. *Giesserei (Duesseldorf)*, vol. 16, no. 41, Oct. 11, 1929, pp. 945-952 and (discussion) 952-953, 4 figs.

Relations between copper and iron alloys are set forth; casting difficulties with bronze; relation between tendency to segregation in raw and cast material; deoxidation of molten bronze with carbon; influence of deoxidation on one segregation, and other causes for zone segregation; new deoxidation process is described.

COPPER-LEAD ALLOYS

The Effect of the So-Called Utah Shale Fluxes on the Dispersion of Lead in Copper-Lead Alloys, G. L. Craig. *Metals and Alloys*, vol. 1, no. 4, Oct. 1929, pp. 162-165, 5 figs.

Use of Utah shales and similar materials from other localities as fluxes in melting metals and alloys is discussed; copper-lead system; properties of shales; experimental procedures; chill, green-sand and dry-sand molds used in preparing casting; examination of castings; under conditions investigated shales used are without effect upon distribution of lead particles in copper-lead mixtures; no beneficial effect of shales noted in any respect.

COPPER-TIN ALLOYS

The Copper-Tin Alloys, J. L. Haughton. *Metallurgist (Supp. to Engineer Lond.)*, Nov. 29, 1929, pp. 167-168, 2 figs.

Review of work of W. Broniewski and B. Hackiewicz describing their own work, commencing with methods of preparation of samples for study of electrical resistance, thermo-electric power, electrolytic potential, dilatation and microstructure. (To be continued.)

Analysis of "Glanzmetall" (Analyse von Glanzmetall), L. Schweitzer. *Chemiker-Zeitung (Koethen)*, vol. 53, no. 47, June 12, 1929, pp. 457-459.

Glanzmetall is name given to some bright alloys which are often made by working up metal wastes; they differ from usual copper alloys by containing more lead and from common tin alloys by containing more copper.

COPPER-ZINC ALLOYS

P.M.G. A New Alloy. *Iron and Coal Trades Rev. (Lond.)*, vol. 119, no. 3219, Nov. 8, 1929, p. 715.

Brief description of substitute for gunmetal, phosphor bronze, etc.; alloy retained formula 88:10:2, but 10 per cent of tin is replaced by some cheaper metal, composition of which will be secret until publication of patent specification; data on casting properties and mechanical tests.

DIE CASTING ALLOYS

ALUMINUM. Effect of Pores on Strength of Die Castings (Der Einfluss von Poren auf die Festigkeit von Spritzguss), L. Frommer, W. Kuntze and G. Sachs. *V.D.I. Zeit. (Berlin)*, vol. 73, no. 45, Nov. 9, 1929, pp. 1609-1612, 4 figs.

Report on X-ray study and mechanical tests of aluminum die castings, made at German Bureau for Materials Testing and Kaiser Wilhelm Institute for Physical Chemistry and Metallurgical Research; authors estimate effect of blowholes in reducing specific strength of casting at not much over 10 per cent.

DIES, FORGING

HEAT TREATMENT. American v. British Die-Tool Treatments, J. W. Urquhart. *Mech. World (Lond.)*, vol. 86, nos. 2233 and 2234, Oct. 18 and 25, 1929, pp. 373-374 and 386-387.

Oct. 18: Review of compositions and method of hardening forging dies in England and America; tempering limits of drop hammer die block; composition; forging machine die-block requirements; some hot pressing die steels; chromium-steel die tools difficult to machine; English die steels. Oct. 25: Carburized plain and nickel-steel die blocks; essential carburization conditions; hammer work and normalizing; sock-resisting nickel steels; heat-application points of importance.

ELECTRIC FURNACES

Electrical Manufacturers Can Lead by Example. *Elec. World*, vol. 94, no. 20, Nov. 16, 1929, pp. 990-991, 4 figs.

It has been estimated that potential load and energy consumption of industrial applications of electric heat will eventually equal or exceed corresponding values of present motor load; every manufacturer now engaged in electrical industry should have interest in seeing that electric heat applications are extended as rapidly as possible on economic basis; illustrations showing some installations of electric furnaces selected by General Electric Co.'s production department because economic studies proved them to be advantageous.

ANNEALING. Furnace Cooled by Double Auxiliary, R. E. Cox. *Iron Age*, vol. 124, no. 20, Nov. 14, 1929, pp. 1311-1312, 2 figs.

Description of electric annealing furnace having variable but controllable heating and cooling cycles; air is circulated in built-in pipes; further cooling by air blast through chamber.

Calculation of Furnace for Mixed Brass Annealing. *Elec. World*, vol. 94, no. 22, Nov. 30, 1929, p. 1088.

Calculation assuming brass mill having brass in miscellaneous size coils, bars, etc., to be annealed and pan-type furnace is to be used; assumed are; weight of charge, 8000 lb. of brass; temperature, 1100 deg. fahr.; weight of pans, 850 lb.; total time for proper anneal, 1½ hr.; rating of furnace is calculated.

HEAT TREATING. Electric Furnaces for the Heat Treatment of Steel, C.H.S. Tupholme. *Iron and Steel Industry (Lond.)*, vol. 3, no. 1, Oct. 1929, pp. 11-13, 4 figs.

Description of series of electric reheating furnaces designed by Brown, Boveri and Co.; framework of reheating furnaces with vertical muffle; special charging device de-

igned for large furnaces treating sheet etal, billets, rods and tubes; methods of egulation; pyrometer equipment; reheating r rolling and wire drawing.

The Use of Electricity in the Heat Treatent and Welding of Metals, A. N. Otis nd W. L. Warner. *Mill and Factory Illustrated*, vol. 4, no. 5, Nov. 1929, pp. 36-38 nd 92-94, 7 figs.

Applications of electric heat in metal industries; diversified uses in treatment of etals; new types of electric furnaces for razing steel parts together in manufacture f complicated assemblies; use of electric elding in fabrication of structures, frameork of machines, etc.

Furnaces and Drying Ovens. *Elec. Times (Lond.)*, vol. 76, no. 1985, Nov. 7, 1929, p. 777-778, 4 figs.

Large electric annealing furnace and orced-air circulation drying oven are used n switchgear production at Hebburn-Upon-Tyne.

Carburizing and Heat Treatment, F. W. owe. *Elec. Times (Lond.)*, vol. 76, no. 984, Oct. 31, 1929, pp. 719-721, 3 figs.

Electric furnaces used for carburizing nd heat treatment work are described, with perating expenses and data.

HIGH FREQUENCY. Coreless Induction Furnaces with Particular Regard to pecial Designs (Der eisenlose Induktionsfen unter besonderer Berucksichtigung on Sonderausfuehrungen), M. H. Kraemer. *Giesserei (Duesseldorf)*, vol. 16, no. 47, ov. 22, 1929, pp. 1092-1099, 24 figs.; and discussion), p. 1099.

Development of special melting process ith high frequency current; description of park-gap furnace of C. Lorenz Corp. for aboratory use at Metallurgical laboratory f Berlin Institute of Technology; future rospects of inductive melting.

INDUSTRIAL. Industrial Electric Furaces. *Elec. Rev. (Lond.)*, vol. 105, no. 708 and 2710, Oct. 18 and Nov. 1, 1929, p. 674-676, and 773-774, 9 figs.

Illustrations of notes on selected items rom manufacturers' products, i. e., Birlec otating annular-hearth furnace; Metropoli-an-Vickers induction furnace at Vickers-rmstrong Works; double-ended furnace or precious metals, Electric Furnace Co.; ecfur melting pot; 3.2-kw. "Birlec" pusher-ype furnace; Wild-Barfield carburizing furaces; "E.F.C." high-frequency resistance urnaces.

MELTING. Economy and Efficiency of lectric Furnace Is Demonstrated in Mine oundries, W. E. Lewis. *Eng. and Min. J.*, vol. 128, no. 21, Nov. 23, 1929, pp. 820-821, 1 fig.

Description of typical 'Lectromelt' furnace lant for mine foundry, capacity 1¼ tons er hour; tabular data on operating costs or three months.

RESISTANCE. Electric Resistance Furaces for Iron and Steel Industry (Elekrische Widerstandsoefen fuer die Eisenuettenindustrie), V. Paschkis. *Stahl u. Eisen (Duesseldorf)*, vol. 49, no. 47, Nov. 1929, pp. 1685-1691, 22 figs., and (discussion) pp. 1691-1695.

Description of high and low-temperature urnaces; advantages of resistance type for heating, annealing, melting, and drying, and other metallurgical furnaces.

STEEL MAKING. Electric Iron and Steel Melting Performance, D. T. Waby. *Elec. World*, vol. 94, no. 23, Dec. 7, 1929, p. 1131.

Table from Western Society of Engineers' Hunt Award paper for 1929 showing size of furnace, capacity, voltage, type of lining, hours per day, tons per month, number of charges, kilowatt-hour per ton, and pounds of electrodes per ton.

Problems of Quality in Electric Steel Works (Qualitaetfragen im Elektrostahlwerk), K. von Kerpely. *Giesserei (Berlin)*, vol. 16, no. 44, Nov. 1, 1929, pp. 1030-1035, 5 figs.

Description of one-ton Heroult-Linderberg electric furnace, equipped with 500-kva. 3-phase transformer with automatic hydraulic electrode regulation, system BBC; operating data.

Three-Phase Arc Furnaces (Der Drehstrom-Lichtbogenofen), S. Schey. *Giesserei-Zeitung (Berlin)*, vol. 26, no. 20, Oct. 15, 1929, pp. 583-587, 8 figs.

Different parts of equipment of electric steel furnace are described, based on which, aspects for correct design are discussed; details of reactance choking coils, electrode regulation, and high-voltage switch-gear.

FERROMAGNETIC MATERIALS

Thermoelastic Properties of Ferromagnetic Materials and Molecular Field (Les propriétés thermoélastiques des metaux ferromagnétiques et le champ moléculaire), E. Bauer. *Jl. de Physique et le Radium (Paris)*, vol. 10, no. 10, Oct. 1929, pp. 345-359, 4 figs.

Energy calculation and entropy of utilizable energy as function of temperature and specific volume on basis of Weiss' theory; deviations in compressibility and dilatation of substances of invariable physical and chemical composition; results obtained in experiments with nickel and iron; theoretical analysis and hypothesis.

FERROMAGNETISM

Role of the Conduction Electrons in Ferromagnetism (Die Rolle der Leitungselektronen beim Ferromagnetismus), J. Dorfman, R. Jaanus and I. Kikoin. *Zeit. fuer Physik (Berlin)*, vol. 54, no. 3-4, April 4, 1929, pp. 277-296, 7 figs.

Investigation is directed to question whether, in ferromagnetic metal, conduction electrons or electrons bound to positive ions of metal, play role of elementary magnets.

FURNACES

FORGING. Influence of Characteristics of Steel on Design of Forging Furnaces, W. E. Jominy. *Fuels and Furnaces*, vol. 7, no. 11, Nov. 1929, pp. 1691-1695, 3 figs.

Discussion of various metallurgical factors involved in heating steel for forging; scaling, decarburization, control of temperature and atmosphere; rate of heating and ease of forging. Abstract of paper presented before Am. Gas Assn.

HEAT TREATING. Accurate Heat Treating in the Small Shop, F. H. Colvin. *Am. Mach.*, vol. 71, no. 22, Nov. 28, 1929, pp. 877-878, 3 figs.

Suggestions as to furnaces, handling equipment, and other details of small shop hardening problems are offered; description of two oil-fired furnaces and gas-fired rotary with dumping baskets agitated by crank, which were built for William Schollhorn Co., New Haven, Conn.

INDUSTRIAL—HEAT INSULATION. Insulation of Furnace Walls (Ueber Waerme-Iolierung und Gasdurchlaessigkeit von Ofenmauerwerk), W. Brass. *Sprechsaal (Coburg)*, vol. 62, no. 34, Aug. 22, 1929, pp. 613-617.

Theory of heat conductivity is discussed and it is concluded that furnace walls should be surrounded by material of low conductivity whereby necessity of absolute resistance to gas penetration is emphasized, especially in reference to joints.

MELTING. Note on Molybdenum Wire-Wound Furnaces, C. Sykes. *Metal Industry (Lond.)*, vol. 35, no. 18, Nov. 1, 1929, pp. 415-418, 2 figs.

Description of furnaces designed by author for metallurgical and other research purposes; of various types available for operation over temperature range of 1100 to 1650 degrees Cent., molybdenum-wound furnaces most suitable; furnaces with porous tube as inner lining; by using central core which is non-porous, furnace can be constructed in which work can be carried out in atmosphere independent of that necessary for protection of winding. Read before Refractory Materials Section, Ceramic Soc.

GAS ANALYSIS

Gas Analysis (Gasanalytisches), E. Ott. *Gas u. Wasserfach (Munich)*, vol. 72, no. 35, Aug. 31, 1929, pp. 862-863, 2 figs.

New form of gas analysis apparatus is described which depends on determination of hydrogen by burning over cuprous oxide at 270-290 deg. and methane at bright red heat, giving more accurate results than combustion or explosion methods; determination of carbon monoxide; modified form of apparatus is also described.

GEARS AND GEARING

SURFACE HARDENING. Surface Hardening. *Automotive Engr. (Lond.)*, vol. 19, no. 259, Oct. 1929, pp. 377-378, 5 figs.

Description of mechanically operated process for treatment of gears; surface hardening of steel by means of oxyacetylene flame; Shorter installation in Huddersfield works for spur gears up to 6 ft. in diam. helical gears up to 3 ft. in diam. and bevel gears up to 3 ft. 6 in. in diam., ranging from 7 to 1 diametral pitch; mechanical head, and tank and wheel holding pedestal; process intended to produce fully hardened mantensitic band of sufficient width and depth to increase life of wheels without distortion.

HARDNESS TESTING

BRINELL. Brinell Hardness Determination, R. G. Batson and S. A. Wood. *Iron and Coal Trades Rev. (Lond.)*, vol. 119, no. 3218, Nov. 1, 1929, p. 673; see also *Engineering (Lond.)*, vol. 128, no. 3329, Nov. 1, 1929, p. 575.

Notes on tests of steel balls used for hardness determination; statement of conclusions and recommendations made to British Engineering Standard Assn. Abstract of Dept. of Sci. and Indus. Research Special Report, no. 16.

The Hardness of Steel Balls Used for the Determination of Brinell Hardness Numbers of Materials, R. G. Batson and S. A. Wood. *Dept. Sci. and Indus. Research—Special Report*, no. 16, 1929, 30 pp., 3 figs.

Description of research to enable specifications to be prepared for hardness of steel balls used for Brinell hardness testing; hardness tests using diamond indenting tool and by ball flattening method; crushing strength of balls by three-ball method; determination of hardness numbers of steel specimens of high and uniform hardness when made with balls of different hardness; file tests on balls.

Making Use of Irregular Brinell Impressions, M. G. Corson. *Min. and Met.*, vol. 10, no. 266, Feb. 1929, pp. 78-79, 2 figs.

Brief statement to effect that irregular Brinell impressions indicates peculiar state of distribution of internal forces, complex of which yields hardness values; it means that metal tested is particularly inclined to flow along certain planes; in other words, exhibits planes of weakness; two microphotographs illustrated this phenomena.

HEAT TREATMENT

AIR EFFECTS. Effect of Atmospheres on the Heat Treatment of Metals, E. G. de Coriolis and R. J. Cowan. *Indus. and Eng. Chem.*, vol. 21, no. 12, Dec. 1929, pp. 1164-1168, 1 fig.

Effects of various atmospheres on metals during typical heating operations ranging from 350 to 2500 deg. Fahr.; these effects are shown to be essentially of chemical nature; methods for application of heat are considered; account is given of research work of American Gas Assn. in connection with forging; problem of bright annealing is considered, use of steam for bright annealing of copper is discussed.

HIGH SPEED STEEL

CUTTING. Influence of Composition and Heat Treatment on Properties of High Speed Steel Used as Cutting Tool (Om sammansattningens och varmebehandlingens inverkan pa snabbstalens egenskaper som skarande), G. Wallquist. *Jernkontorets Annaler (Stockholm)*, no. 7, 1929, pp. 305-350, 41 figs.

Results of American, German, and Swedish investigations show that vanadium is best alloy for improving cutting qualities of steel and that difference of 25 deg. Cent. from ideal hardening temperature lowers cutting quality 65-86 per cent; ideal annealing temperatures for chromium-tungsten-vanadium high-speed steels and for cobalt steel are given.

IMPACT TESTING

NOTCHED BAR. Fractures in Steel and Significance of Transition Region of Notch Toughness (Die Brucharten des Stahles und die Bedeutung des Uebergangsgebietes der Kerbzaehigkeit), F. Fettweis. *Stahl und Eisen (Duesseldorf)*, vol.

49, no. 45, Nov. 7, 1929, pp. 1621-1627, and (discussion) 1627-1628, 11 figs.

Transition region follows in all cases same laws; explanation of occurrence of fractures must take into consideration crystalline nature of metal grains; attempts to explain these fractures are not satisfactory because they are also applicable to amorphous bodies and too general; also, they are contrary to facts and not always logical.

IRON, ELECTROLYTIC

The Estelle Process—Electrolytic Reduction of Iron from its Ores, A. Estelle. *Can. Min. Jl. (Gardenvale)*, vol. 50, no. 48, Nov. 29, 1929, pp. 1137-1140.

Reprint of article indexed in Engineering Index, 1922, p. 363, from same journal Oct. 1922; author describes his process for electrolytic reduction of iron from its ores and points out its advantages; editorial note refers to paper by S. J. Lloyd, read before Am. Electrolytic Soc. previously indexed from advance paper.

IRON ALLOYS

SPECIFIC VOLUME. Determination of Specific Volumes of Iron, Nickel, and Iron Alloys in Molten State (Bestaemning av specifika volymer foer jaern, nickel och jaernlegeringar i smaelt tillstand), C. Benedicks and N. Ericsson. *Jernkontorets Annaler (Stockholm)*, no. 9, 1929, pp. 423-490, 31 figs.

Method used by C. Benedicks, D. W. Berlin, and G. Phragmen has been improved and experiments show that specific volume of iron-carbon alloys at constant temperature increases with increasing carbon contents; specific volume at constant temperature of iron-nickel alloys were also found to be larger than estimated by formulas.

IRON-PHOSPHORUS ALLOYS

Iron-Phosphorus-Carbon System (Ueber das System Eise-Phosphor-Kohlenstoff), R. Vogel. *Archiv fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 3, no. 5, Nov. 1929, pp. 369-381, 22 figs.

Results of investigation by H. Gontermann; crystallization of alloys with 10 to 21 per cent phosphorus; saturation limit of iron-phosphorus mixed crystals; diagrams of state of iron-phosphorus and iron-phosphorus-carbon systems; transformation of mixed crystals in solid state.

LOCOMOTIVES

ALLOY STEEL. Alloy Steels Here to Stay. *Ry. Age*, vol. 87, no. 22, Nov. 30, 1929, pp. 1299-1300.

Discussion of increasing use in locomotive construction presenting problems which should be solved by joint railroad research.

MALLEABLE IRON CASTINGS

MANUFACTURE. Malleable Iron Castings Produced in the Cupola (Kupolofen Temperguss), O. Brauer. *Zeit. fuer. die Gesamte Giessereipraxis (Berlin)*, vol. 50, nos. 44, 45 and 46, Nov. 3, 10 and 17, 1929, pp. 373-375, 383-384 and 391-392, 6 figs.

Author seeks to show how good and cheap malleable castings can be produced at short notice and in large quantities, without costly plant reconstruction and additions or entirely new plant.

MANGANESE ALLOYS

MAGNETIC PROPERTIES. The Magnetic Properties of some Compounds of Manganese, L. F. Bates. *London, Edinburgh and Dublin Philosophical Mag. and Jl. of Science (Lond.)*, vol. 8, no. 53, Nov. 1929, pp. 714-732, 11 figs.

Properties of two simple compounds of manganese, namely compound of manganese and phosphorus and compound of manganese and arsenic.

MANGANESE STEEL MANUFACTURE

Details of preparation of manganese steel by adding ferromanganese directly to ordinary steel as soon as charge is fused and alloying under initial slag.

Details of preparation of manganese steel by adding ferromanganese directly to ordinary steel as soon as charge is fused and alloying under initial slag.

MAGNESIUM ALLOYS

TEMPERATURE EFFECT. High Temperature Compression and Extrusion Tests of Metals and Alloys (Les essais à chaud métaux et alliages par compression et par filage), A. Portevin. *Revue de Métallurgie (Paris)*, vol. 26, no. 8, Aug. 1929, pp. 435-443, 9 figs.

Results of tests carried out on magnesium and magnesium alloys at temperatures not exceeding 500 deg. Cent.; extrusion tests were carried out on hydraulic press.

METALLOGRAPHY

Introduction to Metallography (Eine Einfuehrung in die Metallographie). *Zeit. fuer die Gesamte Giessereipraxis (Berlin)*, vol. 50, nos. 41, 42, 43 and 44, Oct. 13, 20, 27 and Nov. 3, 1929, pp. 349-350, 358-360, 366-367 and 375-376, 12 figs.

Metallographic analysis includes micrographic and macrographic analysis of internal structure and character of metals and alloys; evaluation of analyses of fusion and solidification processes, heat treatment, deformations, gas absorption, oxidation, and reduction processes, erosion and corrosion, and stress measurements; each of these is discussed, and brief mention is made of etching process.

METALLURGY

The World Engineering Congress in Japan. *Engineering (Lond.)*, vol. 128, no. 3332, Nov. 22, 1929, pp. 684-686.

Summaries of papers on metallurgy: Power in Metallurgy, R. M. Keeney; Ore Prospecting by Swedish Geo-Electrical Methods, H. Lundberg and K. Sundberg; World's Iron Ore Supply, C. K. Leith; Steel Industry in Japan, S. Hattori; How Swedish Iron Is Made, J. A. Leffler; Structure of Troostite, F. F. Lucas; Industrial Application of Alloy Steels, R. Hadfield; Lowering of Critical Points in Molybdenum Steels, T. Mukakami and T. Takei; Arsenic in Steel, A. Bauer; Effect of Hot Blast on Blast Furnace Efficiency, F. Wuest.

Annual Meeting of Gesellschaft Deutscher Metallhuetten und Bergleute, Sept. 15, 1929, in Munich (Hauptversammlung der Gesellschaft Deutscher Metallhuetten und Bergleute am 15, Halle September 1929 in Muenchen), K. Nugel. *Metall und Erz*, vol. 26, no. 17, Sept. 1, 1929, pp. 421-425.

Account of meeting of Society of German Mining and Metallurgical Engineers; business report; reports of different professional divisions on chemistry ore treatment, rolling-mill practice, geophysical exploration, engineering education, research work, etc.; brief abstracts of papers read.

RESEARCH. A Review of Recent Metallurgical Investigations. *Fuels and Furnaces*, vol. 7, no. 2, Nov. 1929, pp. 1717-1720.

Reports of investigations on physical chemistry of a steel making, case carburizing, preparation of iron-carbon-manganese alloys, preparation of pure manganese and constitution of high-chromium steel.

METALS

COLD WORKING. Effects of Cold Working on Physical Properties of Metals, R. L. Templin. *Am. Inst. Min. and Met. Engrs.—Tech. Pub.*, no. 238, 1929, pp. 1-15 and (discussion) 15-17, 2 figs.

Factors of variations in conditions present in remounting, casting, chilling and preheating of initial ingot are assumed to remain constant; factors that must be considered simultaneously with cold working; definition of cold work and equivalent cold work; graphs, tabular data, and formulas are given. Bibliography.

CONTACT POTENTIALS. Contact Potentials Between the Same Metals (Ueber Kontakt-potentiale zwischen gleichen Metallen), W. Ende. *Physikalische Zeit. (Leipzig)*, vol. 30, no. 15, Aug. 1, 1929, pp. 477-480, 2 figs.

Contact potentials between nickel-nickel, platinum-platinum and brass-brass were measured; surfaces of metal plates were treated with acids or by filing, sand blasting, etc., before making determinations; contact potential is conditioned by surface layers and, by suitable treatment of latter, can be done away with for most part.

ELECTRON ABSORPTION. Selective Absorption and Reflection of Slow Electrons on Metals (Ueber selektive Absorption und Reflexion langsamer Elektronen an Metallen), E. Rupp. *Zeit. fuer Physik (Berlin)*, vol. 58, nos. 3 and 4, Oct. 24, 1929, pp. 145-164, 14 figs.

Communication of Research Institute of A. E. G., Berlin; tests concerning electron deflection on thin metal films are discussed; measurable permeability of these films for slow atoms has been found at 40 to 4 volts; it is shown that different metals have maximum of absorption for electrons of certain speed.

GAS CONTENT. The Absorption of Gases by Metals, A. Sieverts. *Metallurgist (Supp. to Engineer, Lond.)*, Nov. 1929, pp. 168-172, 10 figs.

Critical review of present knowledge of solution of gases in liquid and solid metals; details of number of gas-metal systems are discussed. and results represented graphically by two types of curves: isobars, where gas absorbed at constant pressure is plotted as function of temperature, and isotherms where absorption is plotted in terms of pressure, temperature being constant. Abstract of paper read before Deutsche Gesellschaft fuer Metallkunde in 1928.

GRANULATION. The Granulation of Metals, E. A. Smith. *Metal Industry (Lond.)*, vol. 35, no. 17, Oct. 25, 1929, pp. 387-390, 4 figs.

Particular property of metal that enables them to be split up into granules when poured into water is discussed; antiquity of granulation; modern granulated practice; bean and feathered shot; securing uniform product; drying precautions; conditions to be observed for successful granulation.

HARD FACING. Development and Application of Hard-Facing. M. C. Smith. *Machy. (N. Y.)*, vol. 36, no. 3, Nov. 1929, pp. 219-222, 6 figs.

Description of materials and methods employed in newly developed process of applying veneer of hard alloy to metal surface by fusion welding; repairing oil-well tools by welding on metal lead to development of hard-facing; three classes of alloys used in hard-facing; problem of finding suitable hard-facing alloy; theory of mixed-alloy welding rod; substituting hard-facing for heat treatment; kinds of hard-facing alloys available; advantages offered by hard-facing process.

HEAT OF FORMATION. Contributions to Electrochemistry of Iron, Manganese and Nickel (Betraege zur Thermochemie des Eisens, Mangans und Nickels), W. A. Roth. *Archiv fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 3, no. 5, Nov. 1929, pp. 339-346; see also *Zeit. fuer Angewandte Chemie (Berlin)*, vol. 42, no. 41, Oct. 12, 1929, pp. 981-984.

Series of tests were carried out in calorimetric bomb to redetermine heats of formation and dissociation of oxides, sulphides, carbides, and other compounds involved in metallurgical processes; in most cases, observed heats of formation were higher than commonly accepted values; tables of heats of formation are given for compounds of iron, manganese, and nickel.

MAGNETIC — ANALYSIS. Recording Apparatus for Determination of Magnetic Transition Points of Small Samples (Ueber einen Registrierapparat zur Bestimmung magnetischer Umwandlungspunkte an keinen Problem), E. Lehrer. *Zeit. fuer Technische Physik (Leipzig)*, vol. 10, no. 5, 1929, pp. 177-185, 14 figs.

Small sample in quartz bulb is suspended in non-homogeneous field (edge-shaped interpolar space), ponderomotive force being recorded by system of levers and mirror to give ordinate of graph; theory and possibilities of instruments is discussed; curves are shown for iron at various field strengths.

MAGNETIC PROPERTIES. Magnetic Chemistry of Diamagnetic and Paramagnetic Metals and Alloys (Magnétochemie der dia- und paramagnetischen Metalle und Legierungen), H. J. Seemann. *Zeit. fuer Technische Physik (Leipzig)*, no. 10, 1929, pp. 399-408, 4 figs.

General review of modern theories; measurements of susceptibility; relation of susceptibility curve to conditional curve; susceptibility and electric conductivity; sus-

ceptibility and magnetic structure in one-phase transformation, etc.

PASSIVITY. The Theory of Passivity (Bemerkungen zur Diskussion ueber die Passivtaetstheorie), R. Mueller. *Zeit. fuer Elektrochemie (Berlin)*, vol. 35, no. 7, July 1929, pp. 459-460.

Theories of passivity are reviewed and combination of hydrogen and oxygen theory with allotropic change in metal is advanced as explanation of passivity.

Passivity of Metals (Zur Passivitaetsfrage der Metalle), I. N. Stranski and Z. C. Mutattschiew. *Zeit. fuer Elektrochemie (Berlin)*, vol. 35, no. 7, July 1929, pp. 393-395, 3 figs.

It is shown that solution of a crystal can be retarded merely by hampering solution of a few ions or molecules situated on its surfaces and edges.

PLASTIC DEFORMATION. On the Mechanics of the Plastic State of Metals, A. Nadai. *Am. Soc. Mech. Engrs.—Advance Paper*, no. 31, for mtg. Dec. 2-6, 1929, 15 pp., 23 figs.

Mechanical conditions under which plastic flow in metals occurs; rules of stresses and deformation when stresses reach limit of plasticity; plastic flow with rotational symmetry around axis in cylinders; bar in combined tension and torsion; hollow cylinder subjected to high internal pressure and axial forces; pure bending of bar with rectangular cross-section; pure torsion with circular cross-section; hollow cylinder subjected to internal pressure; methods of finding residual stresses after severe plastic deformation.

SEASON CRACKING. Season Cracking of Metals, I. V. Williams. *Beil Laboratories Rec.*, vol. 8, no. 2, Oct. 1929, pp. 77-79, 2 figs.

Phenomenon of cracking of metal pieces of any kind during use, without any apparent outside cause is discussed; name "season cracking" is probably applied due to similarity of this form of failure to cracking of timber when undergoing seasoning.

X-RAY ANALYSIS. X-Ray Methods for the Determining of Flaws in Construction Materials (Roentgenverfahren zur raeumlichen Ausmessung von Fehlstellen in Werkstoffen), C. Kantner and A. Herr, *V. D. I. Zeit. (Berlin)*, vol. 73, no. 24, June 15, 1929, pp. 811-816, 24 figs.

Principles of densographic and stereometrometric processes giving results of great precision far exceeding that of usual methods; description of special apparatus and method of using it, with special reference to study of flaws in castings and welds.

METALS ANALYSIS

Select Methods of Metallurgical Analysis, W. A. Naish and J. E. Clennell, with Introduction by H. C. H. Carpenter. London, Chapman and Hall, 1929, 495 pp., 30 shillings, net. Reviewed in *Metallurgist (Supp. to Engineer, Lond.)*, Oct. 25, 1929, p. 160.

Analytical treatment of rare elements is included both in general scheme for qualitative analysis and for purpose of exact quantitative estimation; last chapter deals with application of spectroscope, and especially of modern ultra-violet spectrograph, to analytical purposes.

Select Methods of Metallurgical Analysis, W. A. Nash and J. E. Clennel with Introduction by H. C. Carpenter. Lond. Chapman and Hall. 30 shillings. Reviewed in *Times Trade and Eng. Supp. (Lond.)*, vol. 25, no. 593, Nov. 16, 1929, p. 215.

For every professional metallurgist, assayer, analytical chemist, there is constant need for compendium of such methods, which can serve as standard work of reference; authors of this book have prepared it with object of satisfying this need for whole field of metallurgical analysis.

SPECTROSCOPIC. Quantitative Spectroscopic Analysis of Metals and Alloys. *Metallurgist (Supp. to Engineer, Lond.)*, Oct. 25, 1929, pp. 149-150.

Notes on use of spectroscope in qualitative analysis; attempts at quantitative work; relative advantages in use of arc and of spark sources.

METALS COATING

The Metals Coating Process and Foundry Practice. *Foundry Trade Jl. (Lond.)*, vol. 41, no. 689, Oct. 31, 1929, pp. 311 and 314, 5 figs.

Description of method of coating metals with Metalayer; process consists of automatically feeding into pistol any mental or alloy obtainable in form of wire, where contact is made with oxyacetylene flame which melts metal wire, and in association with compressed air blows atomized metal against surface to be coated at velocity of about 3000 ft. per sec.; operating details; foundry applications.

METALS CORROSION

The Corrosion of Metals and Its Prevention, H. E. Yerbury. *Instn. Mech. Engrs.—Proc. (Lond.)*, no. 3, May 1929, pp. 545-548.

Discussion covers theories of corrosion, effects on iron, steel and nonferrous alloys, specific instances of corrosion and erosion, preservation of iron and steel and nonferrous alloys under working conditions.

Corrosion of Metals, W. Palmer. *Ingeniors Vetenskaps Akademein (Stockholm)*, no. 93, 1929, 347 pp., 111 figs. partly on supp. plates.

General theory and original tests are described; results show conclusively that law of mass action cannot be used for interpreting dissolving of metals, but theory of local galvanic elements must be accepted; three cases of catalytic action are explained and probably for first time it has been possible fully to explain and even calculate beforehand change of reaction velocity with temperature. (In English.)

METALS CUTTING

Investigation as to Cutting Speeds for Swedish Materials, E. A. Forsberg. *Ingenoirs Vetenskaps Akademien (Stockholm)*, no. 95, 1929, 108 pp.

Results of investigations and experimentation which are described, make new formulas for standard cutting speeds much simpler and shows better agreement with observations than Taylor's formula; Taylor's form-

ula for connection between cutting speed and time of wear found to be very accurate for steel whereas substitute exponent used corresponds better with observations for cast iron. (In English.)

METALS FATIGUE

Fatigue Failure of Machine Parts and Testing Methods for Determination of Durability (Ueber Dauerbrueche an Maschinenteilen und geeignete Prufverfahren zur Bestimmung der Dauerfestigkeit), E. Franke. *Automobiltechnische Zeit. (Berlin)*, vol. 32, nos. 29 and 30, Oct. 20 and 31, 1929, pp. 643-645 and 676-678, 9 figs.

Causes for cracks in machine parts subjected to alternation of stresses are discussed; Woehler's testing method; Krupp impact testing machine for determination of resistance of materials; more recently, a dynamic bending machine with rotating test rod, made by Fa. Schenck, has come into use; torsion testing machine of same firm and of Fa. Losenhausenwerk are also described.

Endurance and Fatigue of Metals (Dauer und Ermuedungsfestigkeit), E. Trebesius. *Maschinen Konstrukteur (Leipzig)*, vol. 62, no. 20, Oct. 15, 1929, pp. 459-460.

General outline of fatigue and endurance properties in metal machine parts; author emphasizes that designer should take proper cognizance of phenomena under discussion, and by adequate proportioning change natural vibration of parts in order to avoid critical super-position of frequencies.

The Fatigue of Metals. *Engineer (Lond.)*, vol. 149, no. 3855, Nov. 29, 1929, p. 584.

According to Gough, in whole experience of National Physical Laboratory, no case has been encountered in which slip bands indicative of plastic yielding have not been apparent at stresses which would never fatigue material; surface defects have considerable influence on lowering fatigue strength; this, taken in conjunction with demonstration that fatigue strength is reduced by contact with liquids which produce chemical effect on metal, leaves little room for doubt that fatigue failure commences at surface of metal.

METALS, MOLTEN

PROPERTIES. Some Factors in Solidification, S. W. Smith. *Foundry Trade Jl. (Lond.)*, vol. 41, no. 690, Nov. 7, 1929, pp. 335-336 and 340.

Factors in solidification of metals which have important results in regard to subsequent behavior of metals when in use; solidification of quiescent metals; three stages in operation of cupellation; behavior of cooling globule; factors in solidification process; removal of oxidizable constituents; liberation of gases; liquid contraction during solidification; molten metals undergoing transfer. Abstract of paper presented before Brit. Inst. of Metals.

METALS RESEARCH

Mitteilungen der Deutschen Materialpruefungsanstalten, Sonderheft 9. Kaiser Wilhelm-Institut fuer Metallforschung und dem Staatlichen Materialprufungsamt zu Berlin-Dahlem. Berlin, Julius Springer, 1929. 149 pp., illus., diagrs., tables. 22,50 r.m.

Twenty reports collected in this volume discuss various structural and mechanical properties of industrial metals; influence of lead on structure of brass; age hardening of silver-aluminum alloys influence of various metals on properties of cast iron, aging of duralumin and variety of problems connected with metal crystals are investigated. Eng. Soc. Lib., N. Y.

METAL SPRAYING

COMPRESSED AIR. Protection of Steel from Corrosion by Non-Ferrous Metals. *Metal Industry (Lond.)*, vol. 35, no. 22, Nov. 29, 1929, p. 518, 2 figs.

Description of mechanical process of spraying molten metal directly on to steel, developed by Metals Coating Co.; wire of any commercial metal is fed at definite rate through nozzle into oxyacetylene or oxyhydrogen flame; molten metal is atomized and sprayed by means of compressed air; special attachment used when tubes are to be coated internally.

Metal Coatings. *Elecn. (Lond.)*, vol. 103, no. 2684, Nov. 8, 1929, p. 560, 5 figs.

Possiblity of applying metallic coatings of thickness from 0.001 in. upward to articles and parts of metallic or non-metallic construction, was recently made practicable by pistol spraying process sponsored by Metals Coating Co.; some electrical applications are discussed.

Metal Spraying by Compressed Air. *Shipbldg. and Shipg. Rec. (Lond.)*, vol. 34, no. 19, Nov. 7, 1929, p. 561, 1 fig., see also *Chem. and Industry (Lond.)*, vol. 48, no. 44, Nov. 1, 1929, p. 1069.

System of metallization can be employed for adding wear-resisting surface to softer material and building up worn or pitted surfaces with same metal; is employed to which is supplied oxygen and acetylene for flame which melts metal to be sprayed, and compressed air, which blows molten metal in finely atomized condition through nozzle of piston on to surface.

Test-Bars Versus Actual Strength, W. J. May. *Mech. World (Lond.)*, vol. 86, no. 2234, Oct. 25, 1929, pp. 385-386.

In using results of tests on specially prepared test bars as forming basis for calculating strength of larger castings or forgings, it is necessary to consider method of preparation of both test piece and finished casting in order to determine how far resulting pieces of metal resemble each other in regard to structure; annealing test pieces often considerably alters test strengths; age of test pieces another influence; melting under different conditions is still another influence.

TENSILE. Tension and Compression. *Metallurgist (Supp. to Engineer, Lond.)*, Oct. 25, 1929, pp. 146-147.

Experience shows that in many ways different mechanical properties of engineering metals are closely linked together; close correlation exists between tensile strength and Brinell hardness, at all events in steel, while fatigue range also seems to be closely related to ultimate tensile strength; closer study of relation between tensile and compression properties of more important structural materials is essential.

NICKEL CORROSION

Properties of Nickel in Caustic Evaporation. R. J. McKay. *Indus. and Eng. Chem.*, vol. 21, no. 12, Dec. 1929, pp. 1283-1287, 7 figs.

In caustic evaporation corrosion rate of nickel in caustic solution is so low that it is negligible; in steam, corrosion may occur with high carbon dioxide content; with control of carbon dioxide content corrosion is minute; potentials as high as 0.5 volt are obtained by contact between nickel and cast iron or steel, and amperages indicating rapid corrosion are obtained in strong hot solutions; serious corrosion will occur only where type of contact and areas exposed favor it.

NITRIDATION

Nitriding Steels—Applications to Automobiles and Aircraft. W. H. Hatfield. *Iron and Coal Trades Rev. (Lond.)*, vol. 119, no. 3219, Nov. 8, 1929, p. 712.

Details of nitridation process; tabular data on analysis of steel and properties of steel after nitriding; high degree of hardness of case produced is reflected in its resistance to wear, and most applications of material will be found in those parts where high resistance to wear is desirable feature. Abstract of paper read before Inst. of Automobile Engrs.

Nitrogen Hardening for Wear Resistance. *Machy. (Lond.)*, vol. 35, no. 888, Oct. 17, 1929, pp. 65-67, 5 figs.

Practice for John Brown and Co., Sheffield, in production and treatment of nitralloy steel is described; hardness of nitrided steel; operation involved in nitriding process.

NONFERROUS CASTINGS

Notes on Castings, L. Shane. *Am. Soc. Naval Engrs.—Jl.*, vol. 41, no. 4, Nov. 1929, pp. 589-596.

Author claims that difficulty in making composition G castings is almost exactly contemporaneous with introduction of open-flame furnace, such as Schwartz, or Rockwell, or Monarch furnaces; writer has solved problem in many places, reducing losses to almost negligible extent by merely requiring manufacturing plants where he has been on duty to return to practices of 35 years ago; procedure in effect at Portsmouth Navy Yard, where difficulties in manufacture of Composition G have been overcome; method of manufacture of manganese bronze; monel castings.

NONFERROUS METALS

RESEARCH. Volume Changes During the Solidification Of Metals and Alloys of Low Melting Point, W. E. Goodrich. *Faraday Soc.—Trans.*, vol. 25, no. 101, Oct. 1929, pp. 531-569, 21 figs.

Metals investigated consist of tin, lead, bismuth, and zinc; lead-tin and bismuth-tin alloys; antimony-tin alloys; lead-antimony alloys; tin-antimony-copper and tin-lead-antimony ternary alloys, and some typical zinc-base alloys containing copper and tin and copper and aluminum.

PIG IRON—PROPERTIES

Oxygen Causes Hard Iron, J. A. Lanigan. *Foundry*, vol. 57, no. 23, Dec. 1, 1929, pp. 1017-1018.

Discussion of results of 20 years' observation and investigation on why two pig irons of same analysis will produce castings differing widely in physical properties; question of what element in pig iron sometimes is present in such quantity as to render iron unfit for foundry use; peculiarities noted in fracture of chill block; thorough knowledge is wanted of effects of various non-metallic elements, particularly oxygen, on cast iron.

PIPE, WROUGHT IRON

WELDING. How to Weld Wrought Iron Pipe. *Iron and Steel of Canada (Gardenvale, Que.)*, vol. 12, no. 10, Oct. 1929, pp. 267-268.

Presence of slag in wrought iron necessitates slight modifications in welding technique; puddle of molten metal should always be maintained at point where rod is being added; with reasonable care during welding operation, impurities can be easily floated to surface; properties of Oxweld No. 1 test welding which rod make it particularly suitable for welding wrought iron.

PRESSURE VESSELS

FORGING. Hollow and Ring Forgings. *Machy. (Lond.)*, vol. 35, no. 890, Oct. 31, 1929, pp. 129-134, 13 figs.

Methods employed by John Brown and Co., Sheffield, in heavy forging and machining cylindrical drums; presses used by Company to make forging in capacity from 2000 to 6000 tons; composition of ingot; trepanning operation; forging process; forging presses; hollow rolling mills; boring drum forgings; turning outside diameter and boring manholes.

PYROMETERS, RADIATION

"Pyropto" a Radiation Pyrometer (Strahlungspytometer Pyropto). *Zeit. fuer die Gesamte Giessereipraxis (Berlin)*, vol. 50, no. 40, Oct. 6, 1929, pp. 343-345, 5 figs.

Description of instrument placed on market by Hartmann and Braun, Frankfurt; it consists of telescope inside of which, between lenses, incandescent lamp is built.

RAILROAD MATERIALS

ALLOY STEEL. Alloy Steels in the Railroad Field, C. McKnight. *Am. Soc. Mech. Engrs.—Advance Paper* no. 26, for mtg. Dec. 2-6, 1929, 7 pp., 9 figs.

Three ranges of steels; carbon steels are used for vast majority of railroad purposes, with alloy steels for special applications, and semi-alloy steels, which are slightly more expensive than carbon steels and with slightly better properties, as compromise; corrosion-resisting materials, nitriding, and miscellaneous metals are discussed and economics of operation are given.

RAILS

TRANSVERSE FISSURES. Transverse Fissures in Rails; Their Cause and Prevention, J. E. Howard. *Eng. News Rec.*, vol. 103, no. 24, Dec. 12, 1929, p. 927.

Abstract of report by engineer-physicist of Interstate Commerce Commission states that transverse fissures in rails are caused

by excessive wheel loads, and that remedy lies in finding some grade of steel which will best endure strains.

REFRACTORY MATERIALS

Refractory Materials (Angriffswirkungen auf feuerfeste Baustoffe), M. Pulfrich. *Korrosion und Metallschutz (Leipzig)*, vol. 5, no. 9, Sept. 1929, pp. 193-199; see also brief abstract in *Chem. and Met. Eng.*, vol. 36, no. 11, Nov. 1929, pp. 691.

Influences which attack refractories in service in processing equipment can all be classified under headings, heat, mechanical and chemical corrosion; extreme cold does not harm refractory wall or lining, or stored shapes, if dry; but frost breakage may be expected if refractory is wet; for some uses, refractories can be protected from erosion and corrosion by coating; e. g., a ceramic glaze or graphite.

STAINLESS STEEL

The Stainless Steels and Irons, L. W. Spring. *Valve World*, vol. 26, no. 11, Nov. 1929, pp. 382-384.

Notes on nickel-chrome alloys, carburized steels, nitralloy, etc.

Stainless Steel (Om s.k. rostfritt stal), E. Gauffin. *Tekniska Foreningens i Finland Forhandlingar (Helsingfors)*, vol. 49, no. 3, Mar. 1929, pp. 48.

Brief outline of specification of stainless steel, its tempering, and heat treatment.

Recent Developments in Corrosion-Resistant and Heat-Resistant Steels, J. T. Mathews. *Indus. and Eng. Chem.*, vol. 21, no. 12, Dec. 1929, pp. 1158-1164, 4 figs.

Discussion of various types of corrosion and heat-resistant steels and comparison of relative merits; results of scaling tests.

STEAM TURBINES

BLADES—METALS. Turbine Blading Steels, R. Hadfield. *Mar. Engr. and Motorship Bldr. (Lond.)*, vol. 52, no. 626, Nov. 1929, pp. 459-460, 3 figs.

Description of steel known under name of Hecla/ATV developed primarily to provide turbine blading material not subject to fissuration by superheated steam; it also possesses quality that it is not subject to embrittlement by intercrystalline penetration of hydrogen and other gases.

STEEL

AGE HARDENING. The Effect of Aging and Blue-Brittleness on the Toughness of Low-Carbon Steel, A. Kuehle. *Metals and Alloys*, vol. 1, no. 4, Oct. 1929, pp. 172-175, 11 figs.

Result of investigation of repeated and single impact toughness of two steels, cold worked and subsequently aged, naturally and artificially; series of studies to determine relative sensitivity of two steels to blue brittleness; contradictory and puzzling results of repeated impact test. Paper translated from *Mitteilungen aus dem Forschungs-Institut der Vereinigte Stahlwerke Aktiengesellschaft Dortmund.*

AIRCRAFT. Steels for Automobiles and Aeroplanes, W. H. Hatfield. *Automobile Engr. (Lond.)*, vol. 19, no. 261, Nov. 1929, pp. 464-474, 1 fig.

Review of steels available for use in manufacture of automobiles and airplanes; manufacture of steel used in automobile and airplane production reviewed; ordinary carbon and alloy steels; valve steels; spring steels and springs; steels for chassis frames and bodies; tubes; nitriding steels and nitralloy; rustless steels; rustless sheets and strips; most suitable steels for each part is recommended in five-page table.

AUTOMOBILE. Steels for Automobiles and Aeroplanes, W. H. Hatfield. *Automobile Engr. (Lond.)*, vol. 19, no. 261, Nov. 1929, pp. 464-474, 1 fig.

Review of steels available for use in manufacture of automobiles and airplanes; manufacture of steel used in automobile and airplane production reviewed; ordinary carbon and alloy steels; valve steels; spring steels and springs; steels for chassis frames and bodies; tubes; nitriding steels and nitralloy; rustless steels; rustless sheets and strips; most suitable steel for each part is recommended in five-page table.

COLD WORKING. Hardness and Crystallization Changes by Annealing Cold-Rolled Steel with 1.15, 0.90 and 0.60 per cent Carbon (Hardhets och strukturforandringar vid glodgning av kallvalsat stal, etc.), R. Jonson. *Jernkontorets' Annaler (Stockholm)*, no. 5, 1929, pp. 207-235, 69 figs.

Recrystallization temperatures were found higher for steel with higher carbon contents; microscopical studies showed no relation between recrystallization and structure changes; annealing prolonged from 1 to 5 hours showed little influence on hardness and structure. Bibliography.

DRILL—TESTING. Testing and Treatment of Hollow Drill Steel at North American Mines (Det ihaliga borrstalets provning och behandling vid Nordamerikanska gruvor), L. Nordenfelt. *Jernkontorets Annaler (Stockholm)*, no. 5, 1929, pp. 236-244, 3 figs.

Block test is described and reliability discussed; forging and tempering of drill steel as done at different mines are described.

MAGNETIC HARDENING. Hardening of Superhardened Steel by Magnetism—Lattice Resonance Hypothesis, E. H. Herbert. *Mech. Eng.*, vol. 51, no. 12, Dec. 1929, p. 949.

Brief abstract of paper read before Iron and Steel Inst., previously indexed from advance paper no. 5, Sept. 1929.

SLAG CONTENT. Method of Slag Determination (En slaggbestamningsmetod och nagra praktiska resultat av densamma), H. Kjerrman. *Jernkontorets Annaler (Stockholm)*, no. 4, 1929, pp. 181-199, 13 figs.

Observations of slag of different kinds in steel; methods for microscopical analysis of slag content; relation between slag contents and strength; investigations of sulphate contents in chromium ball-bearing steel.

TEMPERATURE EFFECT. Factors of Safety at High Temperatures. *Engineer (Lond.)*, vol. 148, no. 3852, Nov. 8, 1929, pp. 497-498.

Modern demand for apparatus to stand high stresses at high temperatures has introduced considerations which call for re-examination of whole question of factors of safety; for mild steel, such as is used for steam piping, failure will ultimately occur with stress of less than 4 tons per sq. in. if temperature is maintained at 500 deg.

cent.; 450 deg. cent. seems to have proved safe temperature with ordinary materials.

Properties of Steels at High Temperatures (Les qualites des aciers à haute température), C. Galatoire-Malegarie. *Génie Civil (Paris)*, vol. 95, no. 16, Oct. 19, 1929, pp. 373-377, 6 figs.

Review of recent studies of behavior of special steels at high temperatures and high pressures with special reference to equipment of Issy-les-Moulineaux steam-electric plant; special treatment of steels used; results of tests and operating experience; computation of factor of safety.

Working Stresses for Steel at High Temperatures, D. S. Jacobus. *Am. Soc. Mech. Engrs.—Advance Paper*, no. 22, for mtg. Dec. 2 to 6, 1929, 5 pp., 7 figs.

Results of practical experience used for setting maximum working stresses up to temperatures of 800 deg. fahr.; above this, working stresses are scaled down in proportion to falling off of stresses producing creep; first curve of creep points published by J. J. French used and working stresses for given temperature are taken at ⅔ creep-point stress; effect on stress of piercing boiler shell with holes and of transmitting heat through equalization of stress through extension of most-strained fiber.

The Strength of Steels at High Temperatures, Houdremont and Ehmcke. *Metallurgist (Supp. to Engineer, Lond.)*, Oct. 25, 1929, pp. 155-156.

Authors have taken as basis of comparison tensile strength as determined in tests occupying 15 to 20 minutes; considering first lower temperature region, it has been shown that proportionality exists for steels of pearlitic nature between creep strength at 500 deg. cent. and yield point at room temperature.

Creep of Steel under Simple and Compound Stresses, R. W. Bailey. *Engineer (Lond.)*, vol. 148, no. 3853, Nov. 15, 1929, pp. 528-529, 8 figs.

Two series of tests are recorded; one was made upon steel tubes strained separately at same temperature by tension and torsion producing equal maximum shear stress; other was made upon lead pipes under internal pressure with superimposed axial loading. Paper read before World Power Conference, Tokio.

STEEL ANALYSIS

CARBON DETERMINATION. Magnetic Rapid Determination of Carbon in Steel with Carbometer of Melmberg (Die magnetische Schnelbestimmung des Kohlenstoffs im Stahl mit dem Karbometer von Malmberg), P. Klinger and H. Fuche. *Archiv fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 3, no. 5, Nov. 1929, pp. 374-352, 13 figs.

Description of Malmberg apparatus and magnetic determination of carbon; plotting of calibration curves; influence of alloying elements.

STEEL CALORIZING

Calorizing Steel to Prevent Oxidation, B. J. Sayles. *Machy. (N.Y.)*, vol. 36, no. 3, Nov. 1929, pp. 226-228, 4 figs.

Method of applying calorizing process to mild steel to protect it against oxidation and corrosion at high temperatures is described; application of calorizing to steel tubes; construction of calorizing furnace for revolving retorts; time required for calorizing operation.

STEEL CASTINGS

Cold-Shortness in Special Steels. *Iron and Steel Industry (Lond.)*, vol. 3, no. 1, Oct. 1929, p. 29, 1 fig.

Question of whether structure of casting, surrounding gas atmosphere, applied pressure and internal stresses favor decomposition of cementite is discussed; discussion of article by F. Rapatz which appeared in Stahl und Eisen, No. 25, 1929; unevenness of heating has influence on production of cold shortness.

STEEL CORROSION

Corrosion of Copper-Bearing Steel in Air and in Different Waters (Ueber den Rostvorgang gekupferten Stahles an der Atmosphaere und in verschiedenen Waessern), C. Carius and E. H. Schulz. *Archiv. fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 3, no. 5, 1929, pp. 353-358, 8 figs.

Oxidation potential of salt solutions containing ferrous and ferric irons; reactions of copper during corrosion of copper-bearing steel in distilled water, in air, and in diluted aqueous salt solutions; observations of copper in different forms on surface of copper steel during corrosion in air and water and its significance.

STEEL DEOXIDATION

Use of Binary Alloys for Deoxidizing Steel. *Iron and Steel of Canada (Gardenvale, Que.)*, vol. 12, no. 10, Oct. 1929, p. 252.

Materials first studied were most commonly used single deoxidizers such as manganese, aluminum, and silicon; none of binary alloys studied was found to give type of inclusion desired; proper use of two deoxidizers is first to add desired amount of weak deoxidizer, and then, through addition of stronger deoxidizer, inclusion can be formed which contains appreciable amount of each oxide. Abstract of unspecified report on jointed research by U. S. Bur. of Mines, Carnegie Inst. of Technology, and Met. Advisory Board.

STEEL HEAT TREATMENT

ANNEALING. Testing of Tempered Steel (Provning av hardat stal), A. Lundgren. *Jernkontorets Annaler (Stockholm)*, no. 8, 1929, pp. 367-374.

Communication from State Testing Laboratory on investigation of influence of speed of cooling after annealing on properties of tempered steel and influence of speed of annealing on properties of tempered steel; it is shown that increasing time of annealing increases resistance to bending and impact, but decreases hardness.

NORMALIZING. Normalizing. *Metallurgist (Supp. to Engineer, Lond.)*, Nov. 1929, pp. 166-167, 3 figs.

One of most recent developments of heat treatment in steel industry of America has been continuous normalizing in steel in form of uniform, relatively thin, sections such as sheets and strips, of low-carbon steel; advantages of this process and design of furnaces for carrying it out are described.

Abstract of article previously indexed in Heat Treating and Forging under classification of Sheet Steel—Heat Treatment.

STEEL MANUFACTURE

Speed of Reaction and Equilibrium in Steel Melting (En studie oever reaktionshastighet och jaemviktsfoerhallanden vid stalsmaeltning), F. Sandelin. *Jernkontorets Annaler (Stockholm)*, no. 10, 1929, pp. 519-530, 2 figs.

Study made to clarify question of equilibrium between slag and steel; question is treated theoretically and author explains investigation made by himself and discusses results obtained, which show that reaction takes place at moderate speed and not spontaneously as formerly assumed; that concentration of ferrous-oxide is constant and steel may absorb carbon from furnace gases during refining period. Bibliography.

ELECTRIC PROCESS. High Quality Steel is Produced in Electric Furnaces, H. D. Phillips. *Foundry*, vol. 57, no. 23, Dec. 1, 1929, pp. 997-998.

Practice at plant of Empire Steel Casting Co., Reading, Pa. for manufacture of acid steel in electric furnaces is discussed; general method used in manipulation of heats, with special emphasis on deoxidation of slag and removal of practically all of ferrous oxide from slag; 1-in. diameter round gate off green sand casting was bent 90 deg. with sledge hammer.

STEEL METALLOGRAPHY

Atlas Metallographicus. Lief. 3-5; Tafel 17-40, by Hanemann and Schrader. Berlin, Gebrueder Borntraeger, 1929, 3 parts, plates, pt. 3, 6, 60 r.m.; pts. 4 and 5, 6, 75 r.m. each.

The new sections of atlas contain 143 photomicrographs of structure of steel after various heat treatments; photography is good and structures are clearly shown; clear description accompanies each plate. Eng. Soc. Lib., N. Y.

ALPHA VEINING. Sub-boundaries in Ferrite, E. Ammermann and H. Kornfeld. *Metallurgist (Supp. to Engineer, Lond.)*, Oct. 25, 1929, p. 147.

Review of article published in *Stahl u. Eisen*, Aug. 15, 1929; authors give clear proof that no direct connection with cold working or with deformation at blue heat can exist; they demonstrate this by showing that microscopic feature in question occurs quite as freely in steel of suitable composition which has been normalized as in cold-worked material.

STEEL, SILICON

Silicon Steel for Structural Work. *Eng. Progress (Berlin)*, vol. 10, no. 11, Nov. 1929, pp. 305-307, 3 figs.

Tests made conjointly with Berliner Aktiengesellschaft fuer Eisengiesserei und Maschinenfabrikation and with Dresden Technical College led to introduction of silicon steel containing up to 1 per cent of silicon with average manganese content of 0.8 per cent and carbon content varying between 0.12 and 0.2 per cent.

STEEL TESTING

HARDNESS. Recalculation of Hardness Coefficients and Differential Measuring Method for Determining Thickness of Case (Umrechnung von Haerteziffern, etc.), A. Wallichs and H. Schallbroch. *Automobiltechnische Zeit. (Berlin)*, vol. 32, no. 31, Nov. 10, 1929, pp. 712-713, 1 fig.

Results of comparative tests to determine relations of different new hardness-testing values to Brinell number; Herbert's pendulum hardness tester is briefly described and commended and differential method for measuring thickness of case described by E. G. Herbert and P. Whitaker before (British) Iron and Steel Institute is also discussed.

TENSILE. Investigation of Soft Steel Rod with 0.06 Per Cent Carbon Strained to Destruction (Untersuchungen an einem bis zum Bruch gereckten Stab, etc.), E. Franke. *Maschinen-Konstrukteur (Leipzig)*, vol. 62, no. 19, Oct. 1, 1929, pp. 439-440, 3 figs.

In tensile-strength calculations full cross-section is taken into consideration which, author states, is not correct, because deformations under load have to be accounted for; results of tests on rod 100 mm. long and 20 mm. in diam. to determine how much tensile stresses deviate from calculated mean values, and how much elongation, cross-sectional shrinkage, and Brinell hardness increase until breaking point is reached.

STRUCTURAL STEEL TESTING

Test on Large Columns of Structural Alloy Steel. *Iron Age*, vol. 124, no. 20, Nov. 14, 1929, pp. 1300-1302, 6 figs.

Result of tests made by Bridge Department of Port of New York Authority, on steel for bridge over the Hudson River, including tests to destruction of half-size fabricated members; 1,000,000-lb testing machine of U. S. Bureau of Standards used; specifications for three types of steel including carbon, silicon, and manganese steel; elastic limit of columns much below yield point of tension specimens.

TOOL STEEL PROPERTIES

Bending Resistance of Tempered and Annealed Tool Steel (Bojningshallfastheten hos hardat och anlopt verktygsstal), K. Amberg. *Jernkontorets Annaler (Stockholm)*, no. 6, 1929, pp. 280-299, 20 figs.

Investigation embraces pure and alloy steel, which has been divided into carbon steel (0.75 to 1.21 per cent carbon) low-alloy chromium and tungsten steel, high-alloy chromium steel, and high-speed steel; electric resistance, breaking load, and elastic limits for bending and also Rockwell hardness, have been determined.

TOOL STEEL SPECIFICATIONS

Test Code for High-Speed Steel for Turning Tools. *Am. Soc. Mech. Engrs.—Advance Paper*, no. 49, for mtg. Dec. 2-6, 1929, 3 pp., 2 figs.

Report of Sub-Committee D on Properties of Materials of A.S.M.E. Special Research Committee on Cutting of Metals is given; A.S.M.E. tentative test code for high-speed tool steel intended for manufacture of lathe and planer tools; grinding, heat treatment, test forging, test cut and work done by cutting tool are outlined.

February

Measuring
ess of Case
, etc.), A.
Automobil-
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